The Journal of the

INSTITUTE OF METALS

and

METALLURGICAL ABSTRACTS

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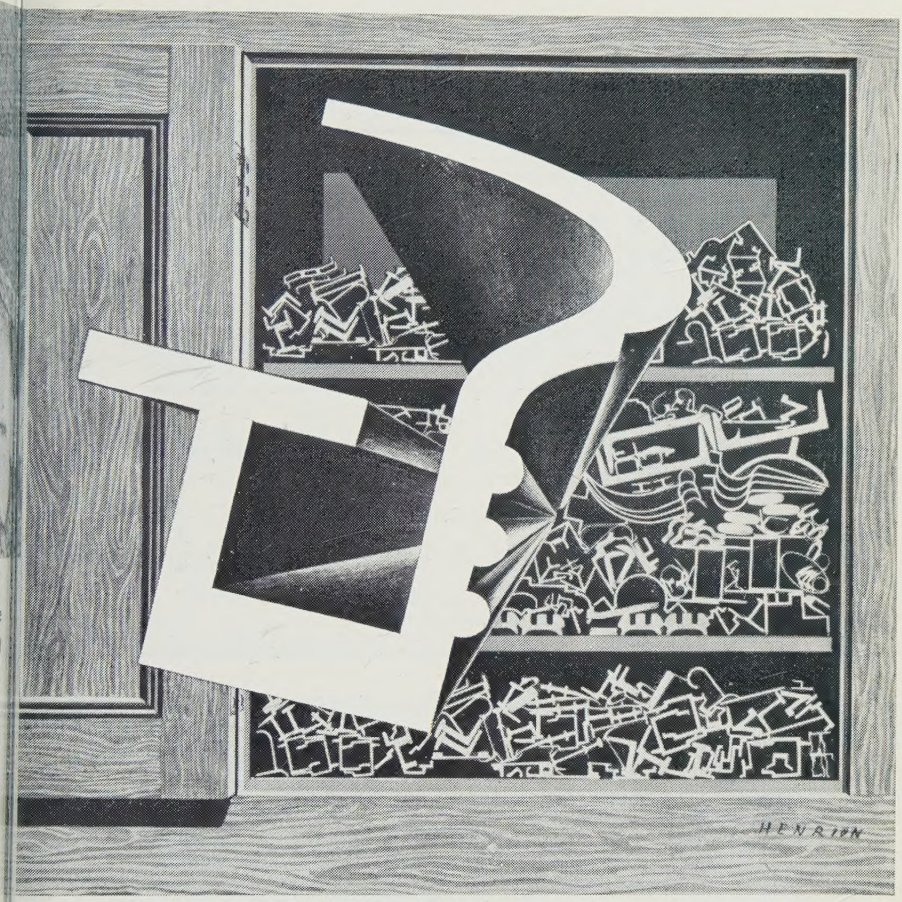

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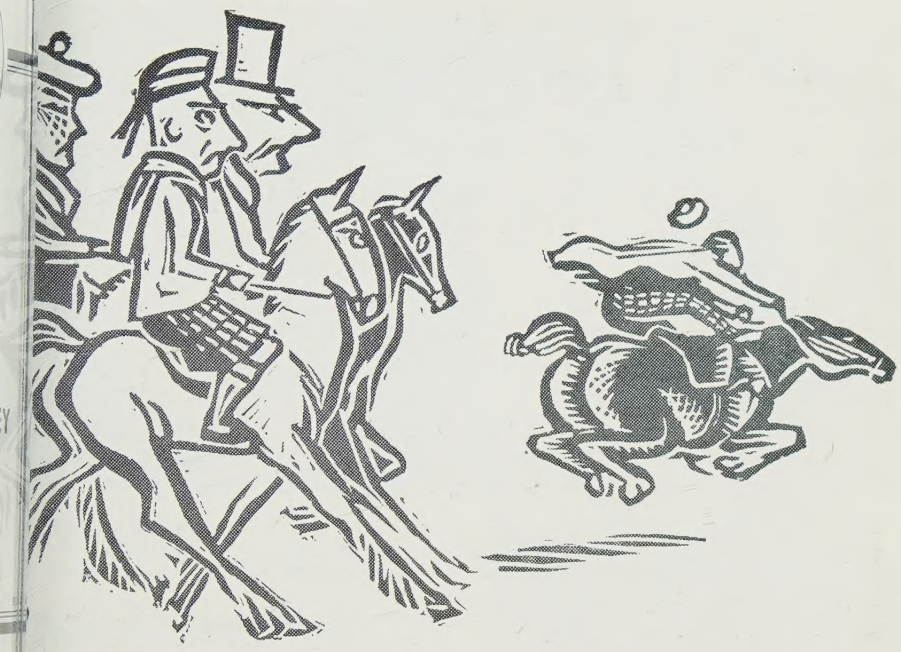

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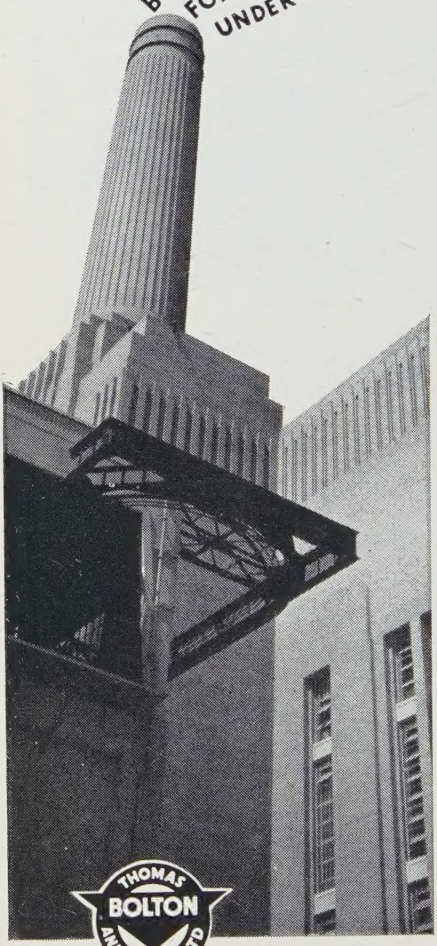

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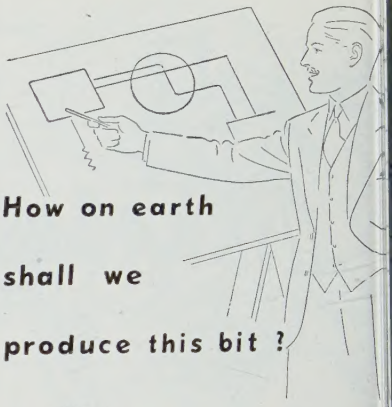

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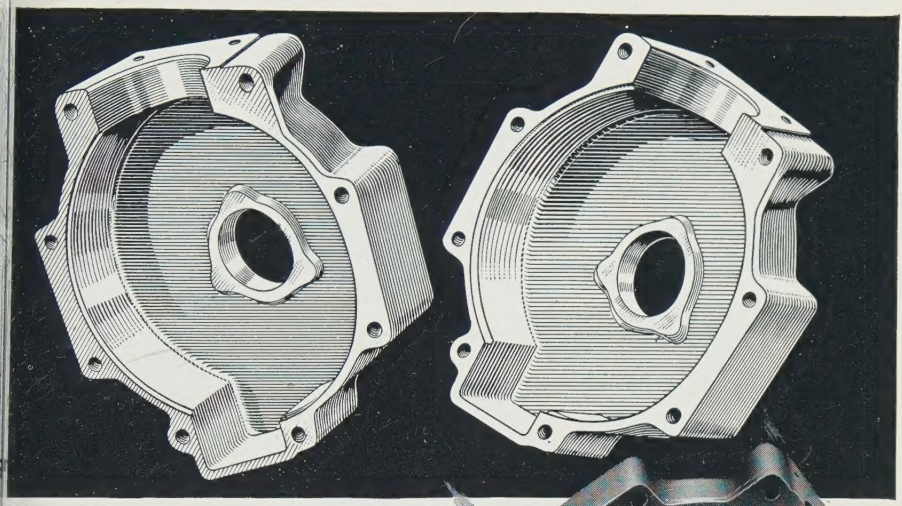

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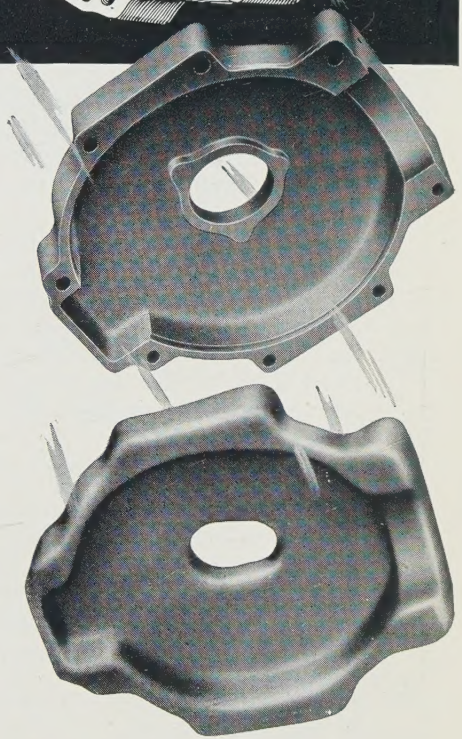

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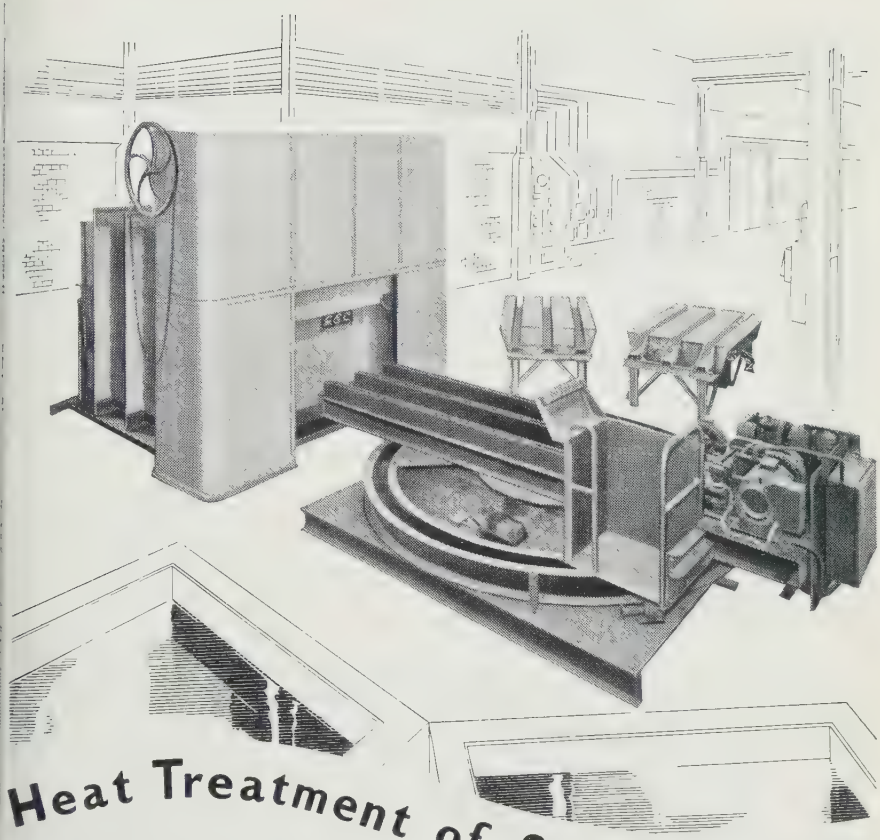

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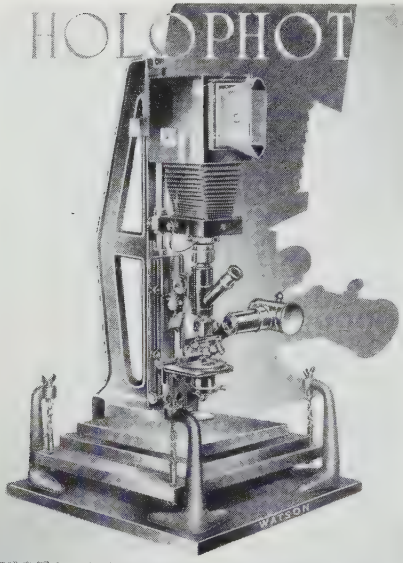

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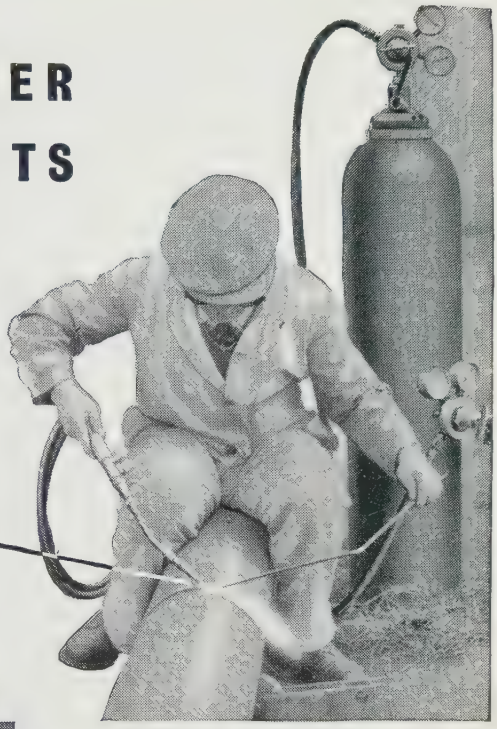

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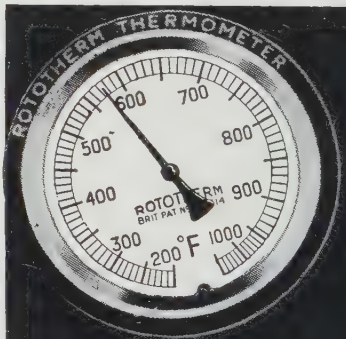

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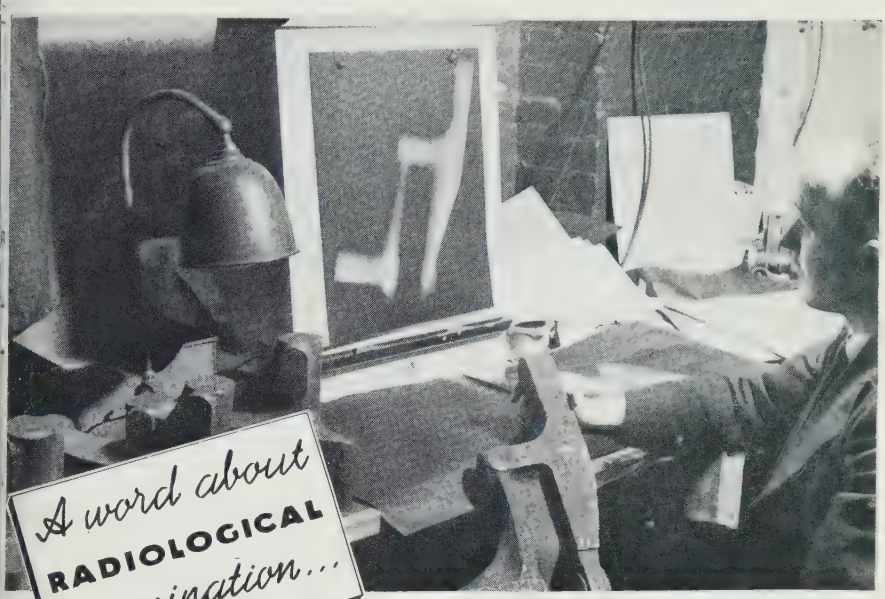

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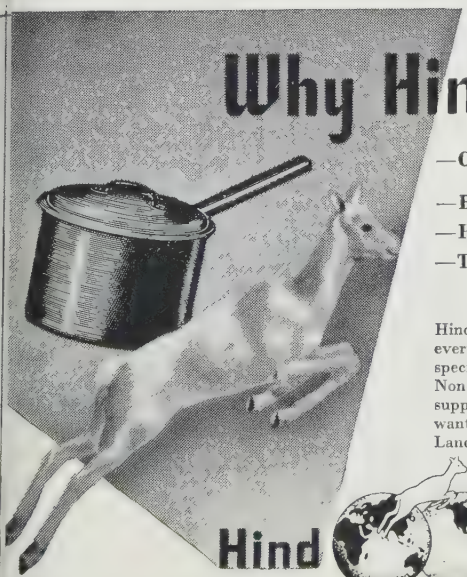

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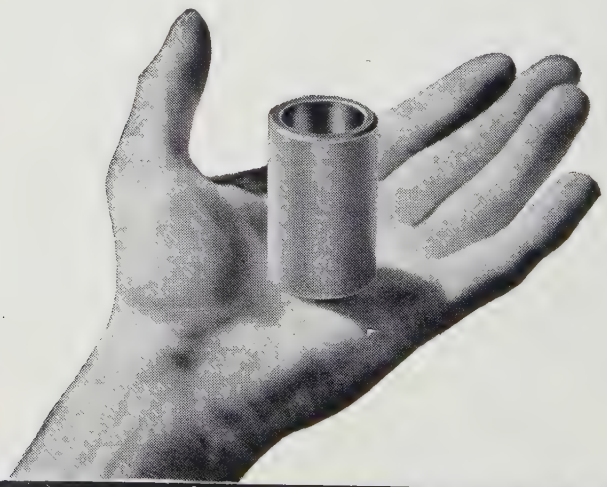

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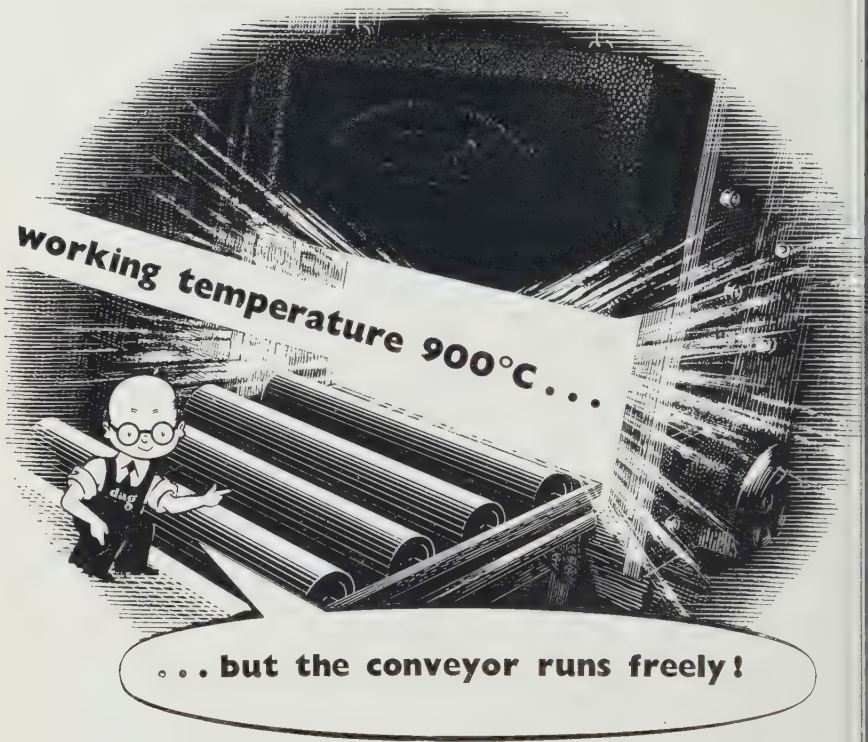

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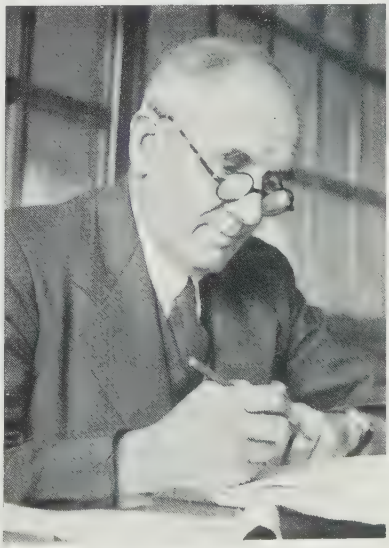
INSTITUTE NEWS AND ANNOUNCEMENTS

PROFESSOR A. O. RANKINE, O.B.E., D.Sc., F.R.S., 1948 MAY LECTURER

Professor Alexander Oliver Rankine was born in Guildford in 1881, and was educated at Guildford Grammar School and University College, London. He graduated with First Class Honours in Physics in 1904, and was thereupon appointed an Assistant in the Physics Department of the same College. He was awarded the London degree of D.Sc. in 1910, chiefly for his research work on the viscosity of gases, and was elected a Fellow of the College in 1912. During the latter part of the first world war he was attached to the Anti-Submarine Division of the Admiralty, as Chief Research Assistant to the late Sir William Bragg, and served in the experimental stations at Harwich and Dartmouth. Besides contributing to the development of submarine detection devices he invented a method of photo-telephony. He was awarded the O.B.E. in 1919.



In the same year he was appointed a Professor of Physics in the Imperial College of Science and Technology and continued his teaching and research work there until 1937. He was elected a Fellow of the Royal Society in 1934. In 1927 he became interested in the subject of Applied Geophysics, especially in relation to the search for oil, and became Advisor on such matters to the Anglo-

Persian (now Anglo-Iranian) Oil Company. He also formed, and took charge of, a section of the Physics Department of the Imperial College to deal with this subject, and trained a number of young physicists to operate geophysical methods in the service of industry. He made many visits abroad—to Persia, Australia, the United States, and Venezuela—to investigate current developments in geophysics. In 1937 he accepted an invitation to become Chief Physicist to the Anglo-Iranian Oil Company, and resigned his post at the Imperial College with the title of Emeritus Professor.

During the second world war Professor Rankine was released by the Company to serve in various capacities in the Admiralty, the Ministry of Aircraft Production, and the Petroleum Warfare Department. His chief work was an experimental investigation of the conditions required for the effective dissipation of fog on airfields by the application of heat—known in its practical operation by the popular name of FIDO. He resumed his work with the Anglo-Iranian Oil Company in 1945 and made further visits abroad—to France, Germany, and Iran—in connection therewith. At the end of 1947 he resigned full-time service with the Company, but continues to act in an advisory capacity.

Professor Rankine has throughout his career devoted much time to the activities of scientific societies. He was Recorder of Section A of the British Association from 1921 to 1924, and President in 1932. From 1923 to 1929 he was Honorary Secretary of the Physical Society, and filled the office of President from 1932 to 1934. He was also Honorary Secretary of the Institute of Physics from 1926 to 1931.

He gave the Royal Institution Children's Lectures on " The Round of the Waters " in 1932–33, and has been a Manager of that Institution on several occasions. In 1945 he was elected Secretary, an office which he still holds.

He is the author of numerous papers published in various scientific journals, principally on the viscosity of gases, the dimensions of molecules, the transmission of speech by light, and applied geophysics.

PERSONAL NOTES

Mr. G. R. Bell, B.Sc., A.R.S.M., A.R.I.C., A.I.M., has taken up the position of General Manager to the Winchester Machine Tool Co., Ltd., 294–296 Lillie Rd., Fulham, London, S.W.6.

Mr. S. C. Britton, M.A., who resigned from the Research Department of the London, Midland, and Scottish Railway, has taken up an appointment with the Tin Research Institute. His address is 30 Boileau Rd., Ealing, London, W.5.

Dr. Ulick R. Evans has received the Willis Rodney Whitney Award in Corrosion Science, of the National Association of Corrosion Engineers (U.S.A.) in recognition of his outstanding contributions to the fundamental knowledge on which the practical control of corrosion is based.

Educated at Marlborough College and King's College, Cambridge, Dr. Evans received a B.A. degree in 1910, an M.A. in 1914, and a Sc.D. in 1932, all from Cambridge University. Immediately after receiving his M.A. degree, world war I started and he served in the British Army from 1914 to 1919. Since

then he has been engaged continuously on scientific work at Cambridge University, mainly in studies of corrosion of metals, thin films on metals, and recrystallization changes in metallic structures. Dr. Evans has published approximately 180 papers, and in addition to lecturing at Cambridge University, he has delivered many lectures by invitation at various centres throughout the world. He is the author of several books on the corrosion of metals which have been published in many different countries. Probably his most important contributions have been his studies on the electrochemical nature of corrosion, and the first quantitative measurements co-ordinating corrosion rates with the amount of electric current flowing during " natural " corrosion were made in his laboratory. Because of the extent, accuracy, and clear presentation of the results of his work, Dr. Evans has had a great influence on the thinking of other workers in the field of corrosion.

Sir William Griffiths, D.Sc., has been awarded the Simms Gold Medal of the Royal Aeronautical Society for his paper on " The Problem of High Temperature Alloys for Gas Turbines ".

Mr. V. C. Hanson has resigned the position of General Manager of Kent Alloys, Ltd., to take up an appointment as a Director of Engineering Reconstruction, Ltd., Consultants, 18 Park St., London, W.1. Mr. Hanson has also been appointed a Director of Scottish Alloys, Ltd., of North Hillington, Glasgow.

Mr. M. R. Hinchcliffe has been appointed a Director of Scottish Alloys, Ltd., Founders and Engineers, Watt Rd., Hillington, Glasgow, S.W.2.

Mr. D. Lane has now joined the Metallurgical Department of Messrs. Newton, Chambers and Co., Ltd., Thorncliffe, near Sheffield.

Mr. W. D. Marsh, B.Sc.Eng., has left Messrs. Murex Welding Processes, Ltd., Waltham Cross, to take up an appointment in the Research Department of Messrs. Hadfields, Ltd., Sheffield.

Professor Robert F. Mehl, Sc.D., Ph.D., arrived in England on 19 April and on 5 May delivered the Third Hatfield Memorial Lecture to the Iron and Steel Institute on " The Decomposition of Austenite by Nucleation and Growth Processes ". He was the guest of the President and Council of the Institute of Metals at dinner at the Savoy Hotel on 13 May. During his stay in Europe, Professor Mehl plans to visit laboratories and works in London, Birmingham, Glasgow, Edinburgh, and Sheffield, and will visit Sweden from 23 to 30 May. He expects to sail for the United States on 2 June.

Professor Mehl was born in Lancaster, Pa., U.S.A., on 30 March 1898. He received his early training at Franklin and Marshall College, Lancaster, Pa., where he graduated as B.S. in 1919; in 1924 he gained the degree of Ph.D. at Princeton University. He was a Proctor Fellow in Chemistry at Princeton University (1922–23); started his professional career as Head of the Department of Chemistry at Juniata College (1923–25), and followed this with a period as a National Research Fellow at Harvard University (1925–27). He next took up the position of Superintendent of the Division of Physical Metallurgy at the Naval Research Laboratory, Washington, D.C. (1927–31), inaugurating this division, and then

became Assistant Director of the Research Laboratories of the American Rolling Mill Company, Middletown, Ohio. In 1932 he became Director of the Metals Research Laboratory and Professor of Metallurgy at the Carnegie Institute of Technology, Pittsburgh, Pa., and in 1935 was made Head of the Department of Metallurgical Engineering in the same Institute; these positions he continues to hold.

Professor Mehl has published about a hundred papers on the constitution of alloys, crystal structure, radiography, the electron microscope, the properties of forging steels, general metallurgy, theory of age-hardening, heat-treatment of steel, &c. He translated Gustav Tammann's book on "The States of Aggregation" (1925) and published his own book on the "Metallurgy of Iron and Steel" (in Portuguese) in Sao Paulo in 1944. His "History of Physical Metallurgy" was issued this year by the A.I.M.E.

Professor Mehl has received many honours. He holds honorary degrees of Franklin and Marshall College (1938), Sao Paulo University (1944), and Stevens Institute of Technology (1944), and, among others, has received the John Scott Medal, awarded by the Directors of the Board of City Trusts, City of Philadelphia (1934), the Howe Medal of the American Society for Metals (1939), and the James Douglas Gold Medal of the American Institute of Mining and Metallurgical Engineers (1945). He has delivered many important lectures, such as the Campbell (1941) and the Sauveur (1940 and 1943) Lectures of the A.S.M., the Annual Lecture of the Institute of Metals Division, A.I.M.E. (1936), and the Priestley Lectures at the Pennsylvania State College (1938).

He is a member of a number of technical and advisory committees associated with the U.S.A. Government and the technical and scientific societies. During the second world war he was an attaché at the U.S. Embassy in London (1945), with the similated rank of Brigadier General.

Dr. B. R. Nijhawan, B.Sc., is now Assistant Director of the National Metallurgical Laboratory, Council for Scientific and Industrial Research, Jamshedpur (via Tatanagar, B.N. Railway), India.

Mr. J. G. Pearce, M.Sc., has been awarded the 1948 E. J. Fox Gold Medal of the Institute of British Foundrymen, in recognition of his services to the foundry industry during the many years in which he has been Director of the British Cast Iron Research Association.

Professor Albert M. Portevin, Member de l'Institut de France, an Honorary Member of the Institute of Metals and of the Iron and Steel Institute, and a Bessemer Medallist of the Iron and Steel Institute, has been promoted to the rank of Commandeur de la Légion d'Honneur. This high distinction is a new recognition of the eminent services which Professor Portevin has rendered to French science and industry. He has contributed to their progress by his teaching work at the École centrale des Arts et Manufactures, at the École supérieure de Fonderie, at the École supérieure de Soudure autogène, &c., by his original researches and his metallurgical discoveries.

Dr. T. L. Rama Char, M.Sc., has been elected to membership of the following scientific institutions: American Institute of

Physics (Associate Member); Electrochemical Society (Member); Indian Institute of Chemical Engineers (Associate Member); and Royal Institute of Chemistry (Associate).

Dr. Cyril Stanley Smith, Honorary Corresponding Member to the Council for the U.S.A., plans to sail for England in the *Queen Mary* on 9 June and will return to the United States on 1 October in the *Queen Elizabeth*. Dr. Smith will attend the conference on the physics of metals to be held in Amsterdam in July and will be present at the Autumn Meeting of the Institute to be held in Cambridge from 14 to 17 September, inclusive. While in England he hopes to visit the principal metallurgical research laboratories.

Sir Arthur Smout, J.P., President of the Institute, has been appointed by the Minister of Education to be a member of the National Advisory Council on Education for Industry and Commerce. This Council is being formed as a result of a recommendation of a Ministry of Education Working Party. The functions of the Council recommended by the Working Party were: (*a*) To keep under continuous review and to advise the Minister on the national policy necessary for the full development of education in relation to industry and commerce, particularly as regards (i) the planning of new developments, including National Colleges; (ii) the expansion of existing facilities; (iii) the improvement of accommodation and equipment; (iv) the development of appropriate research in colleges; (v) the methods of examination and certification of studies; (vi) the provision of scholarships and awards; (vii) the preparation of reports; (viii) the co-ordination of the work of the Regional Advisory Councils and Regional Academic Boards; (ix) other matters relevant to or arising out of the foregoing. (*b*) To maintain contact with industry and commerce and appropriate professional bodies. (*c*) To arrange consultation with the University Grants Committee and other appropriate bodies relating to education of University standard.

Mr. Bernard Thomas, F.Inst.P., has been appointed Chairman and Managing Director of the MacBee Manufacturing Co., Ltd., Bridgnorth, Salop.

BIRTH

Bond-Williams.—On 12 April 1948, to Gwen (née Tomey), wife of Noel Bond-Williams, The Coppice, Lapworth, Warwickshire, a daughter (Susan).

DEATH

The Editor regrets to announce the death of Mr. W. G. Haywood, Managing Director, Haywood Foundries, Ltd., London, on 22 March 1948, his 71st birthday.

LOCAL SECTIONS NEWS

The Local Sections Committee met at the offices of the Institute on 12 March, when, after an informal meeting of the Chairmen and Secretaries, the Committee reviewed the work of the past session and discussed plans for the session 1948–49.

Local Section Committees for the Session 1948–49 have been elected as follows :

LONDON LOCAL SECTION

Chairman :
W. F. RANDALL, B.Sc., A.R.S.M., M.I.E.E., F.I.M.

Vice-Chairman : E. A. G. LIDDIARD, M.A., F.I.M.

Honorary Treasurer : J. D. GROGAN, B.A.

Honorary Secretary :
E. C. RHODES, B.Sc., Ph.D., F.R.I.C., F.I.M.

Ordinary Members :
J. C. CHASTON, B.Sc., Ph.D., A.R.S.M.
R. G. HARPER, M.Sc.
E. C. J. MARSH, B.Sc., A.R.I.C., F.I.M.
W. K. B. MARSHALL, B.Eng., A.I.M.
E. G. V. NEWMAN, B.Sc., A.R.S.M., A.I.M.
C. E. RANSLEY, M.Sc., Ph.D., F.I.M.

Ex-Officio :
G. L. BAILEY, M.Sc., F.I.M. ; A. J. MURPHY, M.Sc., F.I.M. ; J. H. WATSON, M.B.E., M.C., Ph.D., A.R.S.M., F.I.M. (*Past-Chairmen*).

SCOTTISH LOCAL SECTION

Chairman :
A. CRAIG MACDONALD, B.Sc., A.M.I.Mech.E., A.R.I.C., F.I.M.

Vice-Chairman : JOHN ARNOTT, F.R.I.C., F.I.M.

Honorary Treasurer : N. I. MCLEOD.

Honorary Secretary : MATTHEW HAY.

Honorary Auditor : E. H. A. CARLTON.

Ordinary Members :
H. R. BEAUCHAMP. HAROLD BULL.
J. E. CHARD, B.Sc., A.R.S.M., A.R.I.C., A.I.M.
E. G. FLACK. J. GLOVER (*Associate*).
GEORGE MACDONALD, O.B.E., B.Sc.
D. MCSWEIN (*Associate*).

Ex-Officio :
A. B. GRAHAM (*Past-Chairman*).
PROFESSOR G. WESLEY AUSTIN, O.B.E., M.A., M.Sc. (*Past-Chairman*).

QUALIFICATIONS FOR ENTRANCE TO UNIVERSITY SCHOOLS OF METALLURGY

The Joint Committee on Metallurgical Education was formed by the Councils of the Iron and Steel Institute, Institution of Mining and Metallurgy, Institute of British Foundrymen, Institute of Metals, and Institution of Metallurgists, in order to advise and represent them on matters connected with the education of metallurgists.

The Committee has recently considered the standard of general education regarded as desirable for those entering University Schools of Metallurgy in the United Kingdom. The subject is of great importance and the recent publication of the Report of the Secondary School Examinations Council, 1947, makes its consideration at this time very opportune.

The Committee's views and recommendations are set out below. Members are invited to express their opinions on this document, separate printed copies of which may be obtained, on request, from the Secretary, Joint Committee on Metallurgical Education, 4 Grosvenor Gardens, London, S.W.1, to whom all correspondence on this matter should be addressed.

Recommendations

The Joint Committee on Metallurgical Education is concerned with the education of metallurgists for industry and the profession and is representative both of the industries themselves and of those actively engaged in education. Recently it has discussed at some length the education that should have been acquired by those graduating in the various University Schools of Metallurgy in this country, its particular attention having been directed to the general standard of culture attained by the graduates rather than the detailed contents of the University courses. The Committee has chosen this moment to give special attention to this subject, and to express its opinion, because of the recent issue of the White Paper on Examinations in Secondary Schools.* It believes that the Universities will shortly have to determine the conditions under which students will enter their faculties during the next few years; and that all schools will be affected by these decisions.

The Committee appreciates that it is fitted to express an effective opinion in regard to metallurgical education only, though it believes its views on this subject may well be equally applicable to the education of students of other applied sciences, and possibly also to pure scientists. The industrial members of the Committee, themselves employers of scientists trained in different subjects, consider the opinions of the Committee to be applicable equally to graduates in any of the pure or applied sciences. Many voices have been lifted recently to deplore the specialization to which

* Ministry of Education, Report of the Secondary School Examinations Council, 1947. London, 1947: H.M. Stationery Office.

boys and girls too often are subjected during their final years at school, this state of affairs being dictated by the desire that the scholars shall perform as brilliantly as possible in those subjects which carry most weight with the University authorities when entrance is being determined. The Committee joins its voice with these others : it regrets that specialization in the schools has gone to its present lengths ; and it believes that the system produces University students narrow in outlook and interest and lacking in intellectual balance. It considers that it is for the Universities to remedy the present situation by making appropriate adjustments in their entrance requirements. Further, the Committee doubts whether the proposals in the White Paper will of themselves correct all the evils of the present system, though they could do so if the incentive were sufficiently powerful.

The Committee is fully conscious of the differences between one University and another in their entrance requirements, as well as of the difficulties that would attend an attempt to specify a precise and uniform practice. It realizes too that the schools will wish to retain some degree of freedom to act in accordance with their own circumstances, to make full use of their special facilities, and, in certain cases, to preserve old-established traditions. But it feels strongly that, on the one side, the Universities must recognize that it is possible for a man to be a specialist and at the same time to possess a broad culture ; and, on the other, that the schools can, and ought, to provide a suitably broad education without depriving a boy of adequate opportunity to develop in those subjects in which he is most interested. A boy should not, for instance, be permitted to devote his last two years at school exclusively to physics, mathematics, and chemistry, and during these years be effectively divorced from the humanities. If he goes to the University at matriculation level, it is equally inappropriate that the student's work during his first year of residence should be confined to his specialist subjects.

Since University entrance is inevitably accompanied by some measure of competition, the Committee considers that its hopes will not be achieved generally except through explicit action by the Universities themselves. It recognizes that, whereas matriculation, or its equivalent, has always been the minimum standard for entrance, it is very usual for boys to stay at school for a further two years and then take an examination of higher standard—Higher School Certificate or an Intermediate Science—usually in specialized subjects. Boys taking the latter path have been permitted to obtain an honours degree three years later, while those coming up with no higher qualification than matriculation spend a fourth year at the University in order to secure an honours degree. It should not be difficult to ensure that the matriculation test, or its equivalent, is broadly based ; but if a completely balanced education is to be secured it seems that further steps ought to be taken by the Universities. These aim at ensuring that a boy or girl up to what may be called the Intermediate standard should retain a close contact with subjects other than science, and should be asked to provide evidence of competence in such other subjects. To achieve this result in those who have entered at matriculation level is a problem for the Universities to solve, and the Committee believes it can safely be left with them. For school boys seeking to enter at the Intermediate level the

Committee believes that the Universities can act through the entrance examination by demanding that candidates must pass in some subjects chosen from a wide range of non-scientific options.

It may well be objected that the establishment of such an entrance examination will lower the standard set in the science subjects. This is probably true, but the Committee believes, first, that such a lowering would be but a small price to pay for a broader and more balanced culture in metallurgical graduates, and, secondly, that, on the completion of the University course, the difference in the standard attained in metallurgy would not be appreciable. Even if it were, the Committee is so convinced of the need of supplying industry and the profession with graduates well educated in the full as well as in the specialist sense that it would prefer such a result to a continuance of the present system.

It may not be fitting that the Committee should offer detailed suggestions on the entrance examination that the Universities ought to establish, and it recognizes that the Universities can hardly provide a special entrance examination for metallurgists, or even require all candidates to comply with the entrance requirements which the Committee feels should normally be applied. Nevertheless, it wishes to record its opinion that the entrance examinations appropriate for students intending to read metallurgy should recognize that candidates may enter the University either at matriculation level or at that termed the Intermediate in the preceding paragraphs. It is suggested therefore that papers corresponding generally to these two standards should be set in each subject, and that the papers in all the selected subjects should be taken at the same examination. The Committee believes that each examination should contain (1) a paper based on written English, designed to test the student's ability to express himself in writing and to understand the written word, (2) a paper in mathematics, of a standard suited to ensure that the candidate is capable of benefiting from the course he will follow at the University, (3) a paper or papers in chemistry, physics, or general science, and (4) *at least* one paper selected from a number of alternative humanistic subjects other than English. The Committee feels that a call for at least two papers in group (4) would accord more with its wishes, and this would allow a language, other than English, to be made compulsory, the selected language being either classical or modern. In these non-specialist subjects the standard set should be such as would indicate that the candidate is on the way to become a cultured member of society. Success in the papers at the lower standard in all groups would be sufficient to qualify for matriculation entrance only. Candidates wishing to enter at Intermediate standard could profitably take the higher papers in all the subjects chosen, but success at the higher standard should be compulsory in papers (1), (2), and (3), even though the lower standard were accepted in the others. It is thought that examinations framed on these lines would lead boys who sit for University entrance at the age of $17\frac{1}{2}$ to $18\frac{1}{2}$ to read widely in their specialist subjects during their last two years at school and, at the same time, to escape the excessive narrowness too often produced by the existing system.

It remains to refer with some emphasis to a matter not mentioned hitherto, namely, the examinations for University entrance scholarships. Under the system now prevailing, the type of

paper and the competitive nature of the test have probably been the most compelling reasons for that undue and early specialization among school boys which the Committee regrets so intensely and the effects of which it believes to be so unfortunate. If the practice of holding separate scholarship examinations is continued, the Committee believes they must be of the same nature as those suggested for entrance, be based on a broad syllabus and include an adequate proportion of papers in subjects other than those within the candidate's special field. Furthermore, the examination should demand evidence of a proper competence in these other subjects. Unless this step is taken, the benefits deriving from a broadening of the entrance examinations will largely be nullified.

The Committee has debated the selection of students on the basis of a personal interview. This method is now widely used in certain fields, but it seems scarcely practicable to allow it to replace the written test for University purposes. Unless used with the greatest caution the selection of entrants to the Universities in this way might easily result in the exclusion of the scholar of high potentiality, who, at the time he leaves school, is lacking in personality but who may well be admirably suited for research employment in later years. The Committee has no corresponding doubt as to the use of personal interviews when selecting men after graduation.

OTHER NEWS

INSTITUTION OF METALLURGISTS

The Council has appointed Dr. Arthur D. Merriman, G.C., O.B.E., M.A., M.Ed., to the full-time post of Registrar-Secretary, and he took up his duties on 20 April.

Dr. Harold Moore, C.B.E., the Institution's first President, has hitherto served as Registrar, and has agreed to continue to make his experience available in a consultative capacity for a further six months. The Council has recorded its deep appreciation of the assistance that he has given to the Association since its formation.

Under an agreement with the Iron and Steel Institute, Mr. K. Headlam-Morley, Secretary of the Iron and Steel Institute, has acted also as Secretary of the Institution, while clerical duties have been carried out by the staff of the Institute. It has now been decided, by mutual agreement, to bring this arrangement to an end. The Iron and Steel Institute and the Institute of Metals will continue to provide office accommodation for the Institution at No 4 Grosvenor Gardens, London, S.W.1. The Council has expressed its appreciation of the assistance given by the two Institutes and of the services rendered by Mr. Headlam-Morley and his staff.

CAMBRIDGE SUMMER SCHOOL IN METAL PHYSICS, 1948

By the courtesy of Professor Sir Lawrence Bragg, O.B.E., F.R.S., a Summer School in Metal Physics will be held this year in the Cavendish Laboratory.

The School will provide an introduction to the application of physical methods to the examination and utilization of metals, and is intended for those whose researches require a more funda-

mentally physical approach than is usual in ordinary metallurgical and engineering practice. The lectures and demonstrations will deal with (*A*) application of X-ray methods in the examination of metals, (*B*) physical and mechanical properties of metals. Both parts (*A*) and (*B*) will be taken by all attending the School, and for Part (*A*) an elementary knowledge of X-ray diffraction methods and of crystal symmetry will be assumed; during the School it will be possible to provide only a very brief review of these aspects of the subject-matter.

The School will be held from Monday, 23 August, to Friday, 3 September, inclusive. A detailed syllabus and form of application for admission may be obtained from G. F. Hickson, M.A., Secretary of the Board of Extra-Mural Studies, Stuart House, Cambridge, to whom the completed application form should be returned not later than 1 June 1948.

CAMBRIDGE SUMMER SCHOOL IN ELECTRON MICROSCOPY, 1948

By the courtesy of Professor Sir Lawrence Bragg, O.B.E., F.R.S., a Summer School in Electron Microscopy will be held this year in the Cavendish Laboratory, Cambridge.

The School will provide a grounding in the theory and applications of the electron microscope. It is intended for those who have some familiarity with the instrument and who are already, or in the near future will be, operating it in physical, chemical, or biological laboratories. The lectures will deal with the fundamental theory and operating principles of the electron microscope. The various techniques of specimen preparation will be demonstrated, and practised in small groups by the class. The three leading models of instrument will be used, and discussions held on their special characteristics.

The School will be held from 18 to 24 August, inclusive. A detailed syllabus and form of application for admission may be obtained from G. F. Hickson, M.A., Secretary of the Board of Extra-Mural Studies, Stuart House, Cambridge, to whom the completed application form should be returned not later than 12 June 1948.

AMERICAN ELECTROPLATERS' SOCIETY

The Annual Convention and Industrial Finishing Exposition of the American Electroplaters' Society will be held concurrently in Atlantic City from 28 June to 1 July 1948.

The Industrial Finishing Exposition will include displays of equipment and supplies for electroplating, polishing and buffing, cleaning and degreasing, washing and rinsing, lacquering and enamelling. Many pieces of electroplating and other types of equipment will be seen in operation.

CHEMICAL SOCIETY

At the 107th Annual General Meeting, held at Oxford on 18 March, the following officers were elected: President, Sir Ian Heilbron, D.S.O., D.Sc., LL.D., F.R.S.; Treasurer, Sir Wallace Akers, C.B.E., M.A.; Honorary Secretary, Dr. Alexander King, A.R.C.S. Dr. J. L. Simonsen and Professor D. H. Hey will continue to hold office as Honorary Secretaries.

INTERNATIONAL CONGRESS OF INDUSTRIAL CHEMISTRY, 1948

The 21st International Congress of Industrial Chemistry, organized by the Société de Chimie Industrielle, will be held in Brussels from 11 to 19 September 1948.

Full particulars may be obtained from M. Maurice Deschiens, Rapporteur général des Congrès de Chimie Industrielle, Société de Chimie Industrielle, 28 rue Saint-Dominique, Paris 7e.

INTERNATIONAL RHEOLOGICAL CONGRESS, HOLLAND, 1948

The Congress, which will be held at Scheveningen (The Hague) from Tuesday, 21 September, to Friday, 24 September 1948, will be devoted to: (*a*) theoretical problems, (*b*) fundamental experimental methods, (*c*) rheological properties of various systems, (*d*) rheological problems in biology, and (*e*) industrial applications. Plasticity of crystallized materials, soil mechanics, and geophysical problems will be excluded.

Owing to generous support from the Dutch industry and from some Dutch scientific societies, it will be possible to keep the congress fee for participants at f.15 (about £1 8*s*. or $6.50), which will also cover the cost of preprints of papers, but will not include the cost of excursions or the final volume of the Proceedings.

Further particulars may be obtained from the Secretary to the Congress, Mr. R. Houwink, Postbox 66, Delft, Holland.

NATIONAL ASSOCIATION OF CORROSION ENGINEERS, NEW YORK.—1948 CONFERENCE AND EXHIBITION

The 1948 Conference and Exhibition sponsored by the National Association of Corrosion Engineers was held in St. Louis from 5 to 8 April. The four-day session included an extensive technical programme, as well as an All-Corrosion Exhibition.

The technical programme consisted of ten symposia at which approximately forty papers on corrosion, its causes, and methods for combating it, were presented by corrosion authorities. The symposia included: the oil industry, water industry, chemical industry, electrical industry, cathodic protection, protective coatings for metals, general industry, communications, salt-water corrosion, and the gas industry. A general discussion followed each of the meetings.

At the All-Corrosion Exhibition various manufacturers presented displays of the latest materials, methods, and equipment for combating corrosion.

RESEARCH ON THE QUALITY OF ELECTROPLATED COATINGS

The American Electroplaters' Society, through its Research Committee, is engaged in an extensive programme of research on the quality of electroplated coating. This research work was started to help industry to increase the output from existing equipment, reduce the percentage of rejects, and improve the quality of the goods. The programme, which includes eight other related projects, is supported by over 200 industrial concerns, large and small, who look to the American Electroplaters' Society for co-operation and assistance.

Two of its research projects deal with impurities in plating

baths, some of which have a very great effect when present in quantities as small as 1/1000 oz./gall. These studies include development of new and rapid methods for the analysis of impurities and methods for removing them from plating solutions. Even though impurities may not affect the appearance of the plated part when it is new, they may well cause rapid tarnishing, rusting, or other type of failure.

Preliminary results have indicated that the quality of plated products can be greatly improved by good control in the plating plant, and that the ultimate consumer will benefit in longer life of the plated parts.

The research projects are placed at Universities, the National Bureau of Standards, and other research institutions. The effects of impurities are being studied at Michigan State College under Professor D. T. Ewing, and analytical methods are being developed at Lehigh University under Professor E. J. Serfass.

EDUCATION FOR INDUSTRY AND COMMERCE

Mr. George Tomlinson, Minister of Education, announced at the annual meeting of the County Councils Association in London on 7 April 1948, the appointment of Lt.-General Sir Ronald Weeks, K.C.B., C.B.E., D.S.O., M.C., T.D., as first Chairman of the National Advisory Council on Education for Industry and Commerce.

Stating that he was now taking urgent steps to constitute the Council, Mr. Tomlinson said that he had accepted in principle the findings of a Working Party, which recommended in December last the establishment of a National Advisory Council on Education for Industry and Commerce to advise him on the national policy necessary for the full development of education in this sphere. In making its recommendations, the Working Party pointed out that " at this time, as never before, there is a real need for well-trained craftsmen, technicians, technologists, scientists, research workers, and managers ".

Sir Ronald Weeks, who is a graduate of Cambridge University and captained the Cambridge University Football Eleven in 1912, is Deputy Chairman of Vickers, Ltd., Chairman of the English Steel Corporation, Ltd., and a Director of other firms. He was awarded the D.S.O., M.C., and Croix de Guerre in the 1914–18 war, and held various staff appointments in the Home Forces and the War Office during the 1939–45 war, being appointed Director-General of Army Equipment in 1941 and Deputy Chief of the Imperial General Staff in 1942.

DEVELOPMENT COUNCIL FOR JEWELLERY INDUSTRY

The President of the Board of Trade met representatives of both sides of the Jewellery and Silverware Industry on Monday, 22 March, to discuss the establishment of a Development Council for the industry.

It was agreed in principle that such a Council should be set up and its scope and constitution considered. The President promised to circulate to the trade organizations concerned the outlines of a possible D.C. Order for their comments, with a view to more detailed discussion at a future meeting.

GERMAN DIVISION, BOARD OF TRADE: SERVICES TO INDUSTRY

The German Division of the Board of Trade maintains an organization called Technical Information and Documents Unit at 40 Cadogan Square, for the purpose of disseminating the results of technical investigations undertaken in Germany and also giving technical advice on war-time industrial processes in that country. The published C.I.O.S., B.I.O.S., and F.I.A.T. Reports are already well known through their sale by H.M. Stationery Office and the free distribution to libraries and scientific institutions throughout the country. It is not, however, so widely appreciated that in addition to, and in support of, these reports, T.I.D.U. maintains a large library of original German documents, treatises, and research reports. These documents (as distinct from reports) are available for consultation at T.I.D.U. and, moreover, facilities are provided for having copies made in microfilm, microfilm print, or photostat form at the cost of reproduction. Notification of the accession of individual documents to the library is made by the issue of F.D. summaries written in English on yellow foolscap paper. These summaries are sent to trade and research associations and to a limited number of firms whose interests fall outside the coverage of any trade association.

Another aspect of the work of T.I.D.U. which is sometimes overlooked arises from the very close co-operation of the unit with its American counterpart, the Office of Technical Services, Department of Commerce, Washington. As a result of this co-operation T.I.D.U. has most of the material concerning enemy countries announced in the American Bibliography of Scientific and Industrial Reports under P.B. numbers. It is, therefore, advisable to check with T.I.D.U. before ordering from America any document or report bearing a P.B. reference number.

At present a series of reports known as B.I.O.S. Overall Reports is being published. These reports are surveys of the individual industries and attempt to answer the question: "What has Britain to learn from Germany?". This could not be answered until all the available information had been brought over to this country and until British experts had carefully gone through the material and compared it critically with the latest British practice in their own field, because it was felt to be of the utmost importance that this question should be answered as quickly and as fully as possible. The Board of Trade, which is now responsible for B.I.O.S., approached experts in each of the major fields and asked them to reply in the form of critical summaries suitable for publication. They will contain a full bibliography of B.I.O.S., C.I.O.S., F.I.A.T., and J.I.O.A. reports as well as lists of relevant F.D. Summaries.

Requests for review copies should be addressed directly to: Board of Trade, T.I.D.U., Research Section, 40 Cadogan Square, London, S.W.1 (telephone: KENsington 5131, Ext. 149).

RAW MATERIALS GUIDE

The Board of Trade have published a revised Raw Materials Guide which sets out all the raw materials at present controlled by the Board and by the Ministry of Supply. It gives particulars of the relative statutory orders in force, together with information concerning the various changes affecting raw materials which have taken place since October 1946. Details are also given of

the types of control at present operating and the addresses at which enquiries may be made.

Copies of the Guide are obtainable, price 1*s.* 6*d.*, through any bookseller or newsagent, or direct from H.M. Stationery Office, Kingsway, London, W.C.2, and branches.

MINISTRY OF EDUCATION

Technical State Scholarships

Arrangements for the award of Technical State Scholarships will be continued in 1948 upon the same general lines as were announced in 1947.

These Scholarships, to the number of not more than 100, will be offered to students from establishments of further education to enable them to pursue full-time degree courses or courses of equivalent standard at Universities, University Colleges, or establishments of further education. Such courses must normally lead to an Honours Degree or qualification of similar standing. Candidates must have completed a course leading to an Ordinary National Certificate or Diploma or course of equivalent standard, and have been in full time or part time attendance at an establishment of further education for a period of not less than 2 years immediately preceding 1 August 1948, unless the Minister having regard to the special circumstances of the case, allows an exception to be made to this condition. All applications must carry a recommendation for such an award from the Principal of the establishment which the candidate has been attending. Awards will be considered on candidates' performances as finally assessed for National Certificate, Diploma, or Intermediate Degree purposes, or on a record of work certified by the Principal of the establishment. The final selection will be made by a representative Committee.

Candidates must be below the age of 20 on 31 July 1948, although in exceptional circumstances the Minister may extend this age limit to include up to twenty candidates whose claims are the subject of very strong support from the Principals of the establishments of further education at which they are in attendance. In no case, however, will applications be considered from students who would be over the age of 26 on 31 July 1948.

Candidates will be expected to make their own arrangements for acceptance by a University or other institution. In applying for admission to a University or a University College they should make it clear that they are candidates for a Technical State Scholarship. They are strongly advised to satisfy themselves before making application for a Scholarship, both that their qualifications suffice for admission to the course of study they propose to pursue, and that they have secured provisional acceptance by the University or University College of their choice.

Further details of these Scholarships (Form 1. U.T.) and forms of application (Form 2. U.T.) have been distributed to Local Education Authorities and Principals of establishments of further education. Applications must be submitted through the Principal of the establishment concerned to reach the Ministry at Curzon Street House, London, W.1, not later than 30 June 1948.

DIARY FOR JUNE

THURSDAY, 3 JUNE

Royal Society.—E. R. Andrew, M. Desirant, and D. Schoenberg: "The Intermediate State of Superconductors". (Burlington House, Piccadilly, London, W.1, at 4.30 p.m.; tea will be served at 3.45 p.m.)

TUESDAY, 8 JUNE—FRIDAY, 11 JUNE

Institute of British Foundrymen.—Annual Conference. London.

WEDNESDAY, 16 JUNE

Geological Society of London.—Ordinary Evening Meeting. (Burlington House, Piccadilly, London, W.1, at 5 p.m.)

MONDAY, 21 JUNE

Electrodepositors' Technical Society.—H. H. Egginton: "Review of Methods for Testing Thickness of Nickel Coatings"; R. A. F. Hammond: "Jet-Test for Determining Thickness of Lead Deposits". (Northampton Polytechnic, St. John St., Clerkenwell, London, E.C.1, at 5.30 p.m.)

THURSDAY, 24 JUNE

Royal Society.—Sir George Thomson, F.R.S.: "Nuclear Explosions". Bakerian Lecture. (Burlington House, Piccadilly, London, W.1, at 4.30 p.m.; tea will be served at 3.45 p.m.)

WEDNESDAY, 30 JUNE

Institute of Welding.—Annual General Meeting. (Institution of Civil Engineers, Great George St., London, S.W.1, at 2.30 p.m.)

MONOGRAPH No. 3. ATOMIC STRUCTURE FOR STUDENTS OF METALLURGY.

Corrigendum.

Page 11, line 28. *For* "1 Crystal Angstrom Unit = 1 kX. unit = $1{\cdot}00202 \times 10^{-8}$ cm." *read* 1 Angstrom Unit = 10^{-8} cm. = 1·00202 Crystal Angstrom or kX. units."

APPOINTMENTS VACANT

To conform to the requirements of the Control of Engagements Order, 1947, these advertisements are published for the information only of those who are "excepted persons" under the Order.

ASSISTANT METALLURGIST required for British Railways laboratory at Doncaster. Graduate, or A.I.M. or Assoc.Met., with metallographic, iron and brass foundry, and heat-treatment experience. Age not exceeding 35. Salary £250–£470 or £475–£515, according to age, qualifications, and experience. Applications, with details, to Chief Chemist and Metallurgist, Railway Executive, Eastern and North-Eastern Regions, 43 Bridge Terrace, Doncaster, Yorks.

ASSISTANT WORKS MANAGER required for a non-ferrous metal works in London. Applicants should hold a degree in metallurgy, chemistry, physics, or engineering, and should have some previous works' experience. Age should not be over 35 years. Salary £750–£1000 per annum, according to qualifications and experience. Write Box No. 4638, Frost-Smith Advg., 64 Finsbury Pavement, London, E.C.2.

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METALLURGIST required for non-ferrous sand castings, &c. Industrial experience necessary. Age 25–40. Salary according to age and experience, £400 to £800. State age, qualifications, experience, and salary required to Box No. 231, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

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W. S. Angus, Registrar, University Office, 46 North Bailey, Durham.

APPOINTMENTS REQUIRED

GRADUATE METALLURGIST (Member of Institute of Metals), Polish nationality, now in this country after war service, with good experience in metal testing and heat-treatment, and as a University lecturer, seeks appointment research work or industrial testing of non-ferrous metals or steel. Please reply, Box No. 215, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGIST, Ex-Service B.Sc. Student (finals July 1948) seeks position with possibilities for advancement. Free end July. Box No. 232, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

Revised Prices of Institute Publications (Post Free) from 22nd May 1947

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No. 2.—**"Constitution of Alloys Bibliography"**, by J. L. Haughton (1942)	3 6	1 9	2 11
Supplement No. 1 to ditto (1944)	9	5	7
No. 3.—**"Atomic Theory for Students of Metallurgy"**, by W. Hume-Rothery. First reprint (revised) (1947)	7 6	3 9	6 3
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2. **Papers suitable for publication** may be classified as :

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   (*b*) First-class reviews of, or accounts of, progress in a particular field;
   (*c*) Papers descriptive of works methods, or recent developments in metallurgical plant and practice.

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   1. W. Hofmann and W. Jäniche, *Z. Metallkunde*, 1936, **28**, 1 (i.e. year, volume, page).

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THE LATTICE SPACINGS OF THE SOLID SOLUTION OF COPPER IN ALUMINIUM.* 1129

By E. C. ELLWOOD,† B.Sc., A.I.M., Member, and
and (Miss) JEANNE M. SILCOCK,‡ B.Sc.

Synopsis.

The lattice spacings of the solid solution of copper in aluminium have been determined by high-temperature X-ray measurements over the temperature range 400°–600° C. The work was not concerned primarily with the re-determination of the solid-solubility curve of the system, but since micro-examination and room-temperature X-ray measurements involved very little extra work after the preparation of the samples for the high-temperature X-ray work, both methods were used for comparison with the high-temperature results. Experimental difficulties in the measurement of the temperature in the X-ray camera may have been responsible for a lower accuracy than is desirable.

It is shown that the effect of equivalent atomic percentages of copper in distorting the aluminium lattice increases with increasing temperature, owing to the relatively greater differences in the atomic radius of the two elements caused by the difference in thermal expansion. It is also shown that the coefficient of thermal expansion decreases with increasing copper content for similar reasons. Only about 90% of the lattice distortion can be accounted for in terms of size factor.

The solid-solubility curves determined by the three methods agree among themselves and with previous work.

I.—Introduction.

This work is not primarily concerned with the re-determination of the solid-solubility curve of the chosen alloys, but is an attempt to determine, as accurately as possible, the change in lattice spacing with temperature and composition under equilibrium conditions in a single-phase region of an alloy system.

The solid solution of copper in aluminium was chosen for this work because, owing to the technical and scientific importance of the phase field in connection with the heat-treatment and fabrication of the aluminium–copper alloys, much work has been carried out both by classical and X-ray methods on this part of the system. The diagram of Raynor,[1] based on the work of numerous investigators,[2, 3, 4, 5, 6, 7, 8, 9] is accepted as representing the most probable equilibrium conditions.

* Based on a thesis submitted by Jeanne M. Silcock in partial fulfilment of the requirements for the degree of B.Sc., Hons., Applied Science (Metallurgy) of the University of Durham. Manuscript received 30 October 1947.

† Lecturer in Metallurgy, King's College, University of Durham.

‡ Bursar, British Non-Ferrous Metals Research Association; formerly Honours Student, King's College, University of Durham.

II.—Experimental Methods.

1. *Preparation of the Alloys.*

The aluminium was kindly supplied by the British Aluminium Company, Ltd., and had the following analysis: silicon 0·0005, iron 0·0005, copper 0·0005, and aluminium (by difference) 99·9985%. The copper is not a material impurity so far as the present investigation is concerned. The copper was electrolytically refined cathode copper and was presumably of a purity of at least 99·9% copper. Unfortunately, no analysis is available, but since 0·1% total impurity only introduces 0·005% total impurity from the copper in the 5%-copper alloy and 0·001% in the 1% alloy, this is not considered a major omission.

A hardener containing 50% copper was made up by heating together equal weights of copper and aluminium in a fused alumina crucible over a Bunsen burner. The alloy was stirred with a carbon rod before casting into a chill mould and was subsequently used for the preparation of alloys of nominal compositions of 1, 2, 3, 4, 5, and 6% copper by a similar procedure. The alloys were cast into a $\frac{1}{2}$-in.-dia. heated iron mould, since Stockdale [4] found that a very hot mould gave little inverse segregation with these alloys. In spite of this precaution, some inverse segregation was observed in the 4%- and 5%-copper alloys. Reference is made to this below. No attempt was made to test for segregation in the hardener alloy since the compositions of the final alloys were determined by analysis, and while segregation in the hardener may result in departures from the nominal composition, this does not materially affect the results.

2. *Sampling and Analysis of the Alloys.*

The tops and bottoms of the ingots were cropped and the surfaces skimmed off, and turnings taken from the top and bottom of the bars by machining across the whole section. Sections for micro- and X-ray examination were then cut from the top end of the bars. Since analysis by the electrolytic method had shown that there was some discrepancy in the compositions of the top and bottom of the bars and micro-examination showed some inverse segregation in the 5% alloy, it was decided to check the analysis of the actual X-ray specimen. After suitable heat-treatment, the sections for X-ray powder were quenched in water and split across a diameter; and powder was filed from one of the halves as shown in Fig. 1. The remaining half of the section was split to form two pieces, *B* and *C*, to give duplicate samples, so that after filing, three sections remained: *A*, *B*, and *C*.

A obviously contained a preponderance of the outer part of the ingot, while *B* and *C* were representative of the whole cross-section; hence, if appreciable inverse segregation had occurred, *A* would contain more copper than *B* and *C*. The samples obtained in this way were rather small for analysis by the electrolytic method, and instead the copper was determined by the iodine method. The analysis used in plotting the results was the mean of the three determinations on the samples from which the X-ray powder had been taken. In the case of the 5%-copper alloy, in which there was some evidence of inverse segregation, the mean was weighted slightly in favour of the samples which were representative of the whole cross-section of the ingot.

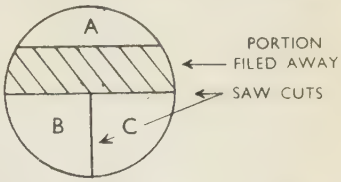

FIG. 1.—Sectioning of the Sample for X-Ray Powder and Chemical Analysis.

The analytical results are shown in Table I.

TABLE I.—*Analytical Results.*

Nominal Composition Copper, wt.-%.	Electrolytic Method.		Iodine Method.			Accepted Analysis.	
	Bottom of Ingot.	Top of Ingot.	*A.*	*B.*	*C.*	wt.-%.	at.-%.
1	1·19	0·96	0·94	0·95	0·99 0·95 *	0·95	0·41
2	2·00	2·02	1·96	1·96	1·90	1·95	0·84
3	2·85	2·81	2·95	...	3·00	2·98	1·28
4	4·04	4·04	4·00	3·97	3·94	3·97	1·73
5	5·09	4·83	5·07	4·97	4·93	4·97	2·19
6	6·08	6·52	...	...	...	...	...

* Repeat.

The 6% alloy was duplex after annealing for a prolonged period at 546° C., and was therefore not analysed accurately. At 550° C., each 0·1 wt.-% copper affects the lattice spacing by about 0·0002 kX. units, which is of the same order as the experimental error in measuring the lattice spacing.

3. *Heat-Treatment.*

All specimens were given a homogenizing treatment in lump form at 546° C.; the time of annealing being two weeks for the 1%-copper alloy, three weeks for the 2% alloy, and four weeks for the remainder. The alloys were sealed in an evacuated hard-glass tube for this treat-

ment and were quenched in cold water at the end of the treatment. All the alloys were free from coring and, with the exception of the 6%-copper alloy, homogeneous. The samples were sectioned and filed as described above, and the filings sieved to pass 100 or 200 mesh. For high-temperature work no further stress-relieving- or heat-treatment was carried out; the specimens were sealed in thin silica tubes and placed in the camera at about 550° C. No evidence was forthcoming that such a procedure gave faulty results. Inevitably, the equilibrium of the specimen is disturbed during the initial heating-up, and precipitation may be caused, but it is thought that holding the specimen for about ½ hr. at 550° C. was sufficient to restore equilibrium and induce stress-relief. It was not possible to follow accurately the temperature of the specimen during heating in the camera, owing to the arrangement of the thermocouple, but the thermocouple temperature only dropped a few degrees when the specimen was put in and quickly returned to its original value.

The preparation of samples for room-temperature X-ray examination was somewhat different from that for those examined at elevated temperatures. For all alloys in the homogeneous field, the samples were prepared from the same powder as was used for the high-temperature work, but after filing, the powder was stress-relieved by annealing at the original homogenizing temperature, namely 546° C., and quenched in water. The lattice spacings of the saturated solid solution were obtained from filings taken from the samples which were used for micro-examination. The filings were stress-relieved at the final annealing temperature, quenched in cold water, and examined. Only micro-samples which were shown to be duplex by micro-examination were used for this part of the work, and hence any possible slight differences in composition from that of the analysed samples were of no account.

Homogeneous alloys were heated and cooled in the X-ray camera as though they were pure metals. The general procedure was to start exposures in the range 540°–560° C., and to proceed to successively lower temperatures until it was estimated that the solid-solubility limit was being approached but not crossed, and then to return to temperatures above 550° C., taking successive exposures at increasing temperature.

In dealing with the 6%-copper alloy, which is duplex at all temperatures, a different procedure had to be adopted. Two specimens were used, one above and the other below the eutectic temperature. Along the solidus, satisfactory results were obtained up to 580° C., i.e. 32° C. above the eutectic temperature and the first appearance

of liquid. No special precautions were taken to reach equilibrium at temperatures along the solidus except to hold the temperature steady for 20 min. before the exposure commenced, and to drop the temperature to about 550° C. over-night. For the solid-solubility curve, however, longer times were allowed to reach equilibrium. These were one hour at 546°, 540°, and 520° C., one day between exposures at 500°–400° C., and for the two points below these temperatures, three days. It is the opinion of the authors that equilibrium is reached fairly rapidly above 400° C., and that the time required for the growth of particles of precipitate of sufficient size to be observed under the microscope is probably considerably longer than that required for the lattice to reach equilibrium.

4. *X-Ray Technique.*

The camera used was of 20-cm. dia. and has been described previously.[10] A modified Van Arkel film arrangement was used, in which diffraction lines with Bragg angles greater than 45° were recorded on a single strip of film 30 cm. long. This arrangement made it necessary to insert supplementary knife-edges so that reference marks were recorded at the ends of the film. The camera was calibrated by direct measurement and the film corrected for shrinkage before calculating the lattice parameters. The $\cos^2\theta$ extrapolation method of Bradley and Jay [11] was used to obtain the true lattice spacings. It is estimated that the possible errors from measurement and calibration were $\pm$ 0·0002 kX. units, and from composition less than $\pm$ 0·0001 kX. units. Errors in temperature measurement are discussed below.

The temperature in the X-ray camera was controlled by hand to $\pm$ 0·5° C. during exposure to X-rays. This variation in temperature does not affect the accuracy of the lattice measurement so long as the mean temperature is recorded. The only ill-effect is a slight broadening of the line, amounting to about 0·05 mm.

5. *Measurement of Temperature.*

The X-ray measurements were made twice. As the work was nearing completion the first time, it was obvious that the eutectic temperature was going to be some 10° or 12° C. too high. This was assumed to be due to some error in temperature measurement, since the melting point of the aluminium–copper eutectic is known fairly accurately. The platinum/platinum–10% rhodium thermocouple, made from 0·4-mm.-dia. wires, was then checked against the melting points of tin, lead, zinc, aluminium, and the aluminium–copper eutectic by the method usually employed for the authors' X-ray camera, which is

to remove the thermocouple from the camera, leaving all leads and the ice-cold junction intact, and calibrate against the potentiometer used for measuring the camera temperature. This check was negative, since the results agreed very closely with earlier calibration figures.

The thermocouple was then calibrated against the melting points of zinc, aluminium, and the aluminium–copper eutectic in the X-ray camera, as described by Wilson,[12] by making up specimens in sealed silica tubes and exposing them at progressively higher temperatures until the diffraction lines disappeared. The eutectic contained a slight excess of copper and exposures were made until the lines from the aluminium solid solution disappeared. The results obtained are shown in Table II, together with the error for a similar calibration at the end of the work.

TABLE II.—*Details of the Calibration of the Thermocouple.*

Metal.	Melting Point, ° C.	Original Calibration in Molten Metal, mV.	Original Calibration in Camera, mV.	Error, ° C.	Error at End of Experiments, ° C.
Zinc . . .	419·5°	3·42	3·45 → 3·46	4·0°	4·5°
Aluminium–Copper Eutectic . .	548·0°	4·655	4·74 → 4·75	12·0°	12·0°
Aluminium . .	660°	5·795	5·88 → 5·89	10·0°	11·0°

Wilson [12] had reported similar but smaller errors previously, but it was naturally very disappointing to find such large errors in temperature measurement. However, since the earlier work had shown that there were no serious anomalies in the system, it was decided to repeat the work, using rather greater temperature intervals than previously, employing the thermocouple calibration obtained in the camera, and making sure that the position of the thermocouple was not altered in any way during the series of experiments. At the end of the experiments, the thermocouple was calibrated in the same manner; the error is shown in Table II.

The final results obtained did not differ very greatly from those for the earlier of the first experiments, but the later results from the first experiments did not agree with the results recorded here. No satisfactory explanation of the discrepancy in calibration has occurred to the authors yet, and the matter is still under investigation.

III.—Discussion of Experimental Results.

1. *General.*

The values of the lattice spacings of the various alloys at elevated temperatures are plotted in Fig. 2, and the values of the lattice spacings

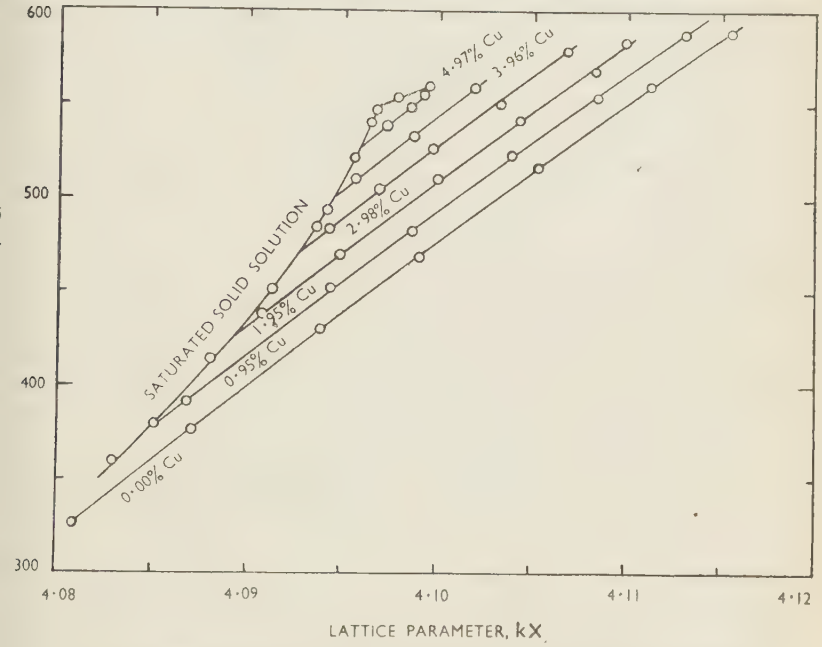

Fig. 2.—Lattice Spacings of the Solid Solution of Copper in Aluminium at Elevated Temperatures.

of quenched alloys are shown in Fig. 3, while the solid-solubility curve obtained from these measurements and the results of micro-examination is given in Fig. 4. The limits of solid solubility as determined by the three methods are in good agreement, although the room-temperature X-ray measurements tend to give results for the solid solubility consistently high by about 0·1% copper.

It has been pointed out by Fink and Freche [13] that the solid-solubility curves for a number of primary aluminium-rich solid solutions, including the solution of copper in aluminium, give rise to a straight line when log S is plotted against $1/T$, where S is the solid solubility in atomic-% at a temperature of $T°$ absolute. In this case, if $1/T$ is extrapolated to the eutectic temperature, the value of S

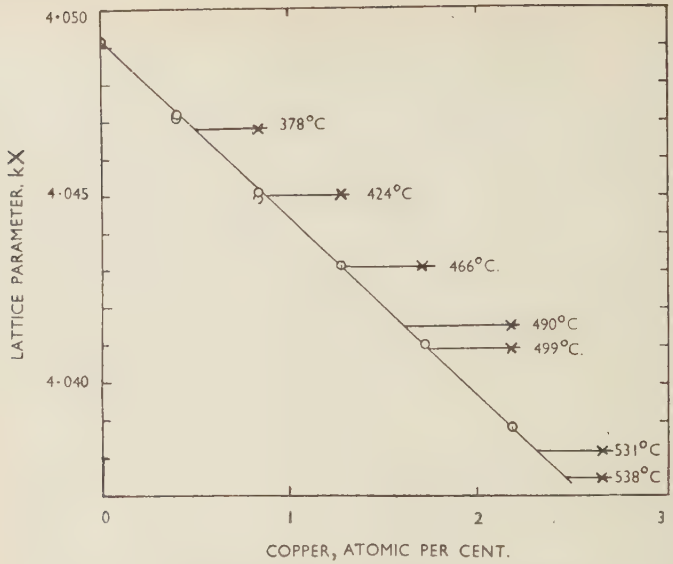

FIG. 3.—Lattice Spacings of the Quenched Samples.

KEY.

○ Alloys quenched from homogeneous field.
× Alloys quenched from duplex field.

Temperatures are those from which the alloys were quenched.

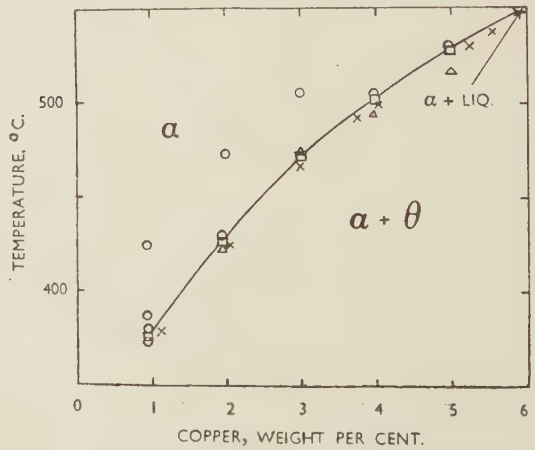

FIG. 4.—Results Obtained by the Three Methods.

KEY.

□ High-temperature X-ray results.
× Room-temperature X-ray results.
○ Micro-examination—homogeneous structure.
△ Micro-examination—duplex structure.

for that temperature is 2·57 at.-% or 5·85 wt.-% copper, which is greater than the value of 5·7 given by Raynor.[1]

2. *The High-Temperature X-Ray Results.*

The curve for the lattice spacing of aluminium confirms the results obtained by Wilson,[14] except at temperatures above 600° C., where the present work gives rather higher values for the lattice spacing. Because of this discrepancy, no measurements taken at temperatures above 600° C. have been recorded.

At 18° C. the value of the lattice spacing of aluminium was determined as 4·0491 kX. units, and at the same temperature the lattice spacing of copper is given by Hume-Rothery [15] as 3·6147 kX. units. At 548° C. the corresponding values are 4·1097 kX. units and 3·6498 kX. units. It is thus to be expected that the solution of copper in aluminium would result in a decrease in lattice spacing. Owing to the relatively greater thermal expansion of aluminium as compared with copper, the difference in spacing of the two elements becomes greater as the temperature is raised, and thus it is to be expected that the change in spacing caused by the addition of unit quantity of copper would increase with temperature, assuming ideal solution. At 18° C., results on quenched specimens show that the lattice spacing of aluminium is decreased by 0·0047 kX. units per atomic-% copper in solution, while at 548° C. (the eutectic temperature) the corresponding decrease in lattice spacing is 0·0051 kX. units. If the solutions are ideal and the only factor affecting lattice spacing is the size of the aluminium and copper atoms, the figures at 18° and 548° C. should be 0·00435 and 0·00459 kX. units, respectively. Thus at room temperature, 93·3%, and at 548° C., 90% of the distortion may be accounted for by the size factor, leaving 7·7 and 10%, respectively, to be accounted for by some other factor, such as valency.

Results obtained in the same way for these and other temperatures are shown in Table III. It is interesting to note that the discrepancy for the quenched samples is of about the same order as that determined at elevated temperatures, in spite of the alloys being supersaturated and having quite different spacings. The differences between the figures shown in column 4 in Table III are probably not significant.

There is no marked change in slope of the lattice-parameter–composition curve at any temperature, probably indicating that the amount of copper added is not sufficient to decrease the electron concentration enough to prevent the overlap of the Brillouin zones in aluminium, and it therefore seems most likely that the discrepancy between the actual lattice distortion and that expected purely from

a consideration of the size factor may be best explained in terms of the electrochemical effect.

It is obvious from the considerations mentioned above that the curves for the thermal expansion of alloys of different composition in

TABLE III.—*Lattice Distortion Produced by* 1 *at.-% Copper at Different Temperatures.*

Temperature, ° C.	Distortion of Lattice per Atomic-% Copper, kX. units.	Theoretical Distortion, Based on Size Factor Only.	Discrepancy, %.
18° *	0·00471	0·00435	7·7
372°	0·00489	0·00448	8·4
428°	0·00489	0·00452	7·5
466°	0·00478	0·00455	5·0
499°	0·00487	0·00457	6·2
528°	0·00501	0·00458	8·6
548°	0·00511	0·00459	10

* Samples quenched from homogeneous field.

the homogeneous phase field will not be identical with that for the thermal expansion of pure aluminium, and that the alloys containing the greatest amount of copper will, in this case, have the lowest coefficient of thermal expansion. The coefficients of thermal expansion at 548° C. were determined from Fig. 2 by measuring the slopes of the curves at that temperature for the five alloys and pure aluminium, and calculating the expansion with reference to the spacing of each alloy at 548° C. The coefficients of thermal expansion determined in this way are shown in Fig. 5, from which it is clearly seen that the coefficient at 548° C. decreases with increasing copper content.

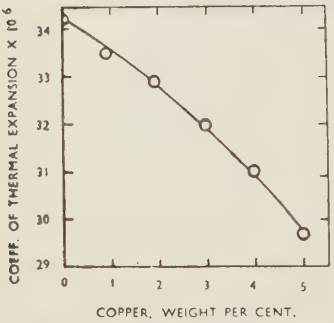

FIG. 5.—Coefficients of Thermal Expansion at 548° C.

If it is required to calculate the lattice spacing of an alloy at elevated temperatures from its spacing at room temperature, it will be seen from the results given above that this can be done only approximately by taking into account the thermal expansion of the solvent and changes in the relative sizes of the solvent and solute atoms at the elevated temperatures. That part of the change in lattice spacing due to effects other than the size factor probably cannot be accurately predicted from low-temperature spacing measurements,

although in this case the effect was relatively constant for alloys quenched from 548° C. and alloys measured at 548° C.

Acknowledgements.

The authors wish to thank Professor C. E. Pearson, and their colleagues in the Metallurgy Department of King's College, University of Durham, for many helpful discussions while the work was in progress.

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FURTHER CORRESPONDENCE ON DR. W. BOAS AND MR. R. W. K. HONEYCOMBE'S PAPER: "THE DEFORMATION OF TIN-BASE BEARING ALLOYS BY HEATING AND COOLING."

(*J. Inst. Metals*, 1947, **73**, pp. 433, 809.)

Mr. Arthur F. Underwood*: The investigation described in this paper is a real contribution to the fundamental knowledge of bearing metallurgy.

Several years ago, a tin-base bearing was returned to us from service for examination because of its uneven surface. Examination showed the crystals to be very large (up to $\frac{1}{4}$ in. across), probably owing to slow cooling after casting. During use in the engine, the crystals had deformed to produce the unevenness, which was readily apparent to the unaided eye. Although this is the only case of distorted crystals brought to our attention, it is interesting that the effect has developed in service operation.

In this country, lead-base Babbitt has nearly replaced tin-base Babbitt in engines for passenger cars. It has been demonstrated that the superiority of lead-base over tin-base Babbitt is more marked as the test temperature is increased. The difference in behaviour of these two metals when subjected to cyclic thermal treatment may be an important factor in the growing use of lead-base Babbitts.

Most bearing tests are conducted by running the engine at constant high speed for long periods of time. It would be very interesting, and worthwhile, to operate an engine fitted with both types of bearings on a schedule which allowed frequent heating and cooling, to assimilate more closely actual operating conditions.

The Authors (*in reply*): We are pleased to have Mr. Underwood's remarks on our paper. It is very gratifying to know that his wide industrial experience agrees with the results obtained in an academic investigation.

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THE EFFECT OF VIBRATION ON A PRECIPITATION-HARDENING ALUMINIUM ALLOY.* 1130

By R. F. HANSTOCK,† Ph.D.

Synopsis.

The dynamic properties of the precipitation-hardening aluminium alloy Hiduminium R.R. 56 have been investigated by determining the relation between damping capacity and strain amplitude for various metallurgical conditions of the alloy. The form of this relation suggests that strain-dependent damping capacity is attributable to dislocations generated under the combined influence of strain and thermal fluctuations. The factors which appear to be mainly responsible for the dynamic behaviour of the alloy are : an internal field of strain, probably originating from foreign atoms, in solution or dispersed in the aluminium lattice, and a strain-concentration factor depending on the degree of precipitation.

Vibration at high strain amplitudes accelerates precipitation and, if sufficiently prolonged, produces changes in metallurgical condition normally associated with temperature ageing and annealing. Being of a directional nature, vibration is probably selective in its action on the grains of a polycrystalline metal and consequently does not affect the average static mechanical properties to a significant extent. However, the grains affected by vibration determine the fatigue properties, and it is shown that the endurance under a given strain amplitude may be inferred from the change in damping capacity during vibration.

I.—Introduction.

The most drastic consequence of subjecting a metal to vibrational strains is fatigue failure, but this can be only the climax of a gradual modification of the properties of the metal. The study of damping capacity is a promising approach to the investigation of pre-fatigue phenomena, for it is well known that this property is very sensitive to the condition of the metal.

Several investigators have already observed changes in the damping capacity of certain metals and alloys subjected to vibrational strains. Read [1] found that the damping capacity of a single crystal of zinc increased during vibration if the strain amplitude was large, while Brick and Phillips,[2] and Lazan [3] observed a decrease in damping capacity during the vibration of some aluminium alloys. Stainless-steel turbine blades, examined by Glikman and Grinberg [4] before and after 12,000 hr. of service, were found to have decreased in damping capacity, although no change could be detected in the mechanical properties of the blades. Annealing restored the damping capacity

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to its original value. They attributed the decrease to either strain-hardening of certain individual grains or dispersed precipitation of intermetallic compounds. Hanstock and Murray [5] found that the aluminium alloy Hiduminium R.R. 56,* heat-treated to a condition of optimum mechanical properties, increased in damping capacity during vibration. Two aluminium alloys containing magnesium, investigated by the author,[6] decreased in damping capacity until fatigue failure occurred.

The damping capacity of a metal does not necessarily arise from a single source. It may be partly of thermo-elastic,[7] magnetic,[8] or plastic origin and may be increased by flaws [9] within the metal. For certain conditions of vibration, the damping capacity of non-magnetic metals is mainly attributable to plasticity and shows a marked dependence on the strain amplitude.

Decrease in damping capacity of plastic origin during vibration indicates that the metal is being strain-hardened, but the reason why an increase is sometimes observed is not immediately apparent. The present work was undertaken to investigate more closely the vibrational behaviour of R.R. 56, a precipitation-hardening alloy, which in certain metallurgical conditions, increases in damping capacity during vibration.

At the start of this investigation, a thorough knowledge of the dependence of the relation between damping capacity and strain amplitude on the metallurgical state of the alloy was considered to be essential for an explanation of the effects of prolonged vibration. Furthermore, if the relation between damping capacity and strain amplitude could be reduced to a mathematical form, preferably derived from current theories of plasticity, it was felt that the problem of identifying the processes activated by vibration might be simplified. The second section of this paper is devoted to the derivation, based substantially on the Becker–Orowan theory of plastic deformation, of an equation relating damping capacity to strain amplitude. The third section presents experimental results demonstrating the validity of this equation, and, later, other observations showing how the damping capacity of the alloy is affected by prolonged vibration. In the discussion, an attempt is made to identify the processes induced by vibration and to indicate their effect on the service performance of the alloy.

II.—Damping Due to Plasticity.

In general terms, Read [1] has referred to the movement of dislocations, of the type considered by Taylor,[10] Polanyi,[11] Orowan,[12] and others, as a source of damping capacity.

* Nominal composition: copper 2, magnesium 0·9, silicon 0·8, iron 0·9, nickel 1·1, titanium 0·1%, aluminium remainder.

An ideal crystal consisting of a perfectly regular array of atoms would require a shear stress of the order $G/2$ (G is the modulus of rigidity) to cause slip between two neighbouring planes of atoms. Normal crystals deform plastically under very much smaller shear stresses and, to account for this, the crystal is assumed to contain dislocations. Briefly, a dislocation is present when one of two neighbouring atom planes contains either one row of atoms more, or one less, than the second plane. In the neighbourhood of the dislocation, the spatial periodicity of the atoms is abnormal, and Taylor has shown that slip between the planes, to the extent of the distance between two neighbouring atoms, can occur by translation of the dislocation. The shear stress required to cause this translation is very much smaller than that required to cause slip in a perfectly regular arrangement of atoms. If new dislocations are generated at a sufficient rate, a macroscopic deformation of the crystal may be produced by a comparatively small shear stress.

According to the Becker–Orowan theory, a dislocation is generated when atoms in a small group are displaced sufficiently from their normal positions, under the combined influence of the applied stress and a thermal fluctuation of the energy of the atoms.

The probability W of the generation of a dislocation by this process is given by Boltzmann's equation:

$$W = A \exp.(-\varepsilon/kT) \quad . \quad . \quad . \quad . \quad . \quad (1)$$

where T is the absolute temperature; k is Boltzmann's constant; A is a constant; and ε represents the accumulation by thermal fluctuation, in a small volume v, of energy sufficient to raise the applied shear stress σ up to the value σ_m which is necessary to generate a dislocation. Thus:

$$\varepsilon = v(\sigma_m - \sigma)^2/2G \quad . \quad . \quad . \quad . \quad . \quad . \quad (2)$$

According to Orowan, the internal shear stress σ must be distinguished from the external stress S which produces the internal stress. Dislocations are considered to be generated mainly in the neighbourhood of abnormalities in the crystal structure, e.g. grain boundaries and foreign atoms. In these regions a stress concentration occurs, so that the stress σ is equal to qS, where q is a stress (or strain) concentration factor. Expressing stresses in terms of corresponding external strains ϕ and ϕ_m, the expression for ε becomes:

$$\varepsilon = \tfrac{1}{2}vGq^2(\phi_m - \phi)^2 \quad . \quad . \quad . \quad . \quad . \quad (3)$$

The quantity $q\phi_m$ now represents the internal strain corresponding to the displacement of an atom at the origin of a dislocation, and, in magnitude, it should be of the order 0·5. The volume v is that in

which a thermal fluctuation must occur to displace a sufficient number of atoms to establish a dislocation. The experimental results given later suggest that if a thermal fluctuation causes a sufficient displacement of *one* atom, the probability of neighbouring atoms being displaced under the prevailing shear stress in such a way as to form a comparatively stable dislocation is very high.

Consequently, the jump of a single atom may be regarded as initiating a dislocation (or possibly an avalanche of dislocations) which grows to stable dimensions and travels under the influence of the prevailing stress. The volume v is, therefore, taken to be equal to that occupied by one atom, i.e. it is given by V/L, where V is the volume per gramme atom of the metal and L is Loschmidt's number.

In order to derive a relation between damping capacity and strain amplitude, it is assumed that the energy ΔE dissipated per cycle is due to translation of dislocations and that it is proportional to the number generated in a given volume. As the number of atoms in a test specimen is very large, the number of dislocations generated will be expressed by the probability W. Assuming that the energy dissipated is also proportional to the total vibrational energy E, the relation between damping capacity $\Delta E/E$ and strain amplitude ϕ is of the form :

$$\Delta E/E = B \exp. [- VGq^2(\phi_m - \phi)^2/2LkT] \quad . \quad . \quad (4)$$

where B is a constant, which for the following reasons may be taken as unity. When $\phi = \phi_m$, dislocations are produced continuously by the applied shear stress, so that plastic flow becomes continuous. For such a condition ΔE will be of the same order of magnitude as E, and consequently no serious error is introduced by assuming that $B = 1$. The quantities ϕ_m and q, appearing in equation (4), express the state of the crystal structure. Their numerical values may be obtained by substituting experimental values of $\Delta E/E$ and ϕ in equation (4), since the other quantities in this equation have the following values :

$V = 10{\cdot}03$ c.c./g. atom (value for aluminium).
$G = 2{\cdot}90 \times 10^{11}$ dynes/cm.^{2} (value for R.R. 56).
$L = 6{\cdot}06 \times 10^{23}$ atoms/g. atom.
$k = 1{\cdot}37 \times 10^{-16}$ ergs/° K.
$T = 293°$ K.

It will be seen that, according to equation (4), the damping capacity due to plasticity is finite even at very low strains. Assuming that $q\phi_m = 0{\cdot}5$, the damping capacity for strains approaching zero will be of the order of 10^{-7} for aluminium alloys. When the strain is increased from zero to ϕ_m, the damping capacity increases slowly at first, but as ϕ approaches ϕ_m the increase becomes very rapid. The equation also

shows that the damping capacity due to plastic flow increases with rising temperature. In this paper, the effect of temperature will not be considered, but it is hoped eventually to determine the temperature function experimentally. Actually, Barnes and Zener [13] have found experimental evidence for a temperature function of the type given by equation (4).

The damping capacity expressed by equation (4) refers to a volume of metal in which the shear strain at any instant has a uniform value ϕ. For various reasons stated elsewhere, the torsional mode of vibration is preferred for the experimental determination of damping capacity. In this mode, the shear strain at any instant varies linearly from zero at the vibration axis to a maximum value at the surface of the cylindrical specimen. Consequently, the observed damping capacity is an average for the non-uniformly strained specimen, and in order to subject equation (4) to experimental verification it is necessary to correct the observed damping capacity for the effect of this non-uniform strain. For torsional tests on cylindrical specimens,[14] the specific damping capacity P_ϕ for a uniform strain ϕ is given by :

$$P_\phi = P + \tfrac{1}{4}\phi \frac{dP}{d\phi} \quad . \quad . \quad . \quad . \quad . \quad . \quad (5)$$

where P is the observed average damping capacity corresponding to a shear-strain amplitude ϕ at the surface of the cylinder.

Measured values of damping capacity always include some damping due to leakage of vibrational energy from the system and may include damping arising from thermo-elastic effects. Since the damping capacity due to plasticity may be expected to be very small (of the order 10^{-7}) at low strains, the damping observed at low strain amplitudes is probably mainly of non-plastic origin. To obtain data for comparison with equation (4), the value of the damping capacity found at the lowest strains (10^{-4}) at which measurements are made is subtracted from all values of damping capacity observed at higher strain amplitudes. Generally, the total damping capacity at a strain amplitude of 10^{-4} is 5×10^{-5} for R.R. 56, and, of this amount, 2×10^{-5} is attributable to external losses of energy.

III.—Experimental.

Full details of the method employed for measuring the damping capacity of specimens having a torsional resonance frequency of approximately 1 kc./s. have been given in earlier papers (*loc. cit.*).

In this investigation, the specimens were prepared from swaged bars of $1\frac{1}{2}$ in. dia., all originating from the same melt.

1. *Heat-Treatment of Specimens.*

The heat-treatments were chosen to bring the alloy to conditions known to be either of practical importance or of interest with respect to the precipitation process. They were as follows:

(1) Solution heat-treatment at 530° C. for 2 hr. From this temperature the alloy is quenched, normally in water at temperatures within the range 20°–100° C., according to the size of the component and the degree to which freedom from residual stresses is necessary. Three temperatures of quenching were adopted, namely 20°, 70°, and 100° C., and specimens so treated are referred to, respectively, as being in the *ST* 20, *ST* 70, and *ST* 100 states. After quenching, precipitation from solid solution proceeds at room temperature so that it is necessary to regard time after quenching as a factor affecting the metallurgical condition established by solution heat-treatment and quenching.

(2) Precipitation at elevated temperatures. Good mechanical properties are attained when the solution heat-treated alloy is maintained at a temperature of 170° C. for several hours. A standard period of 16 hr. was adopted for this treatment, specimens subjected to these thermal conditions being referred to as *ST* 20*A*, &c.

Specimens with pronounced precipitation were obtained by annealing at 360° C. for 2 hr. followed by cooling in the furnace at a rate of 10° C./hr. Specimens in this condition are referred to by the abbreviation *An*–360.

(3) Specimens were also prepared under the conditions specified in specification D.T.D. 246A for large components required substantially free from residual stresses. The processing comprises solution heat-treatment and air cooling, followed by ageing at 200° C. When in this condition the specimens are identified by the code *ST–AN*.

Representative values for the ultimate shear strength of the alloy in the comparatively stable conditions *ST* 20*A*, *ST–AN*, and *An*–360 are, respectively, 18·7, 13·7, and 7·3 tons/in.^{2} Wöhler fatigue-stress limits at 20×10^6 cycles are 9·75 and 8·55 tons/in.^{2} for the condition *ST* 20*A* and *ST–AN* respectively. Assuming a value of 10^7 lb./in.^{2} for Young's modulus, these fatigue stresses correspond to strains of $21{\cdot}8 \times 10^{-4}$ and $19{\cdot}2 \times 10^{-4}$. In the present investigation, vibrational strains of this order of magnitude were produced.

2. *Variation of Specific Damping Capacity with Shear Strain.*

The object of the first series of experiments was to determine the relation between damping capacity and strain amplitude for the alloy

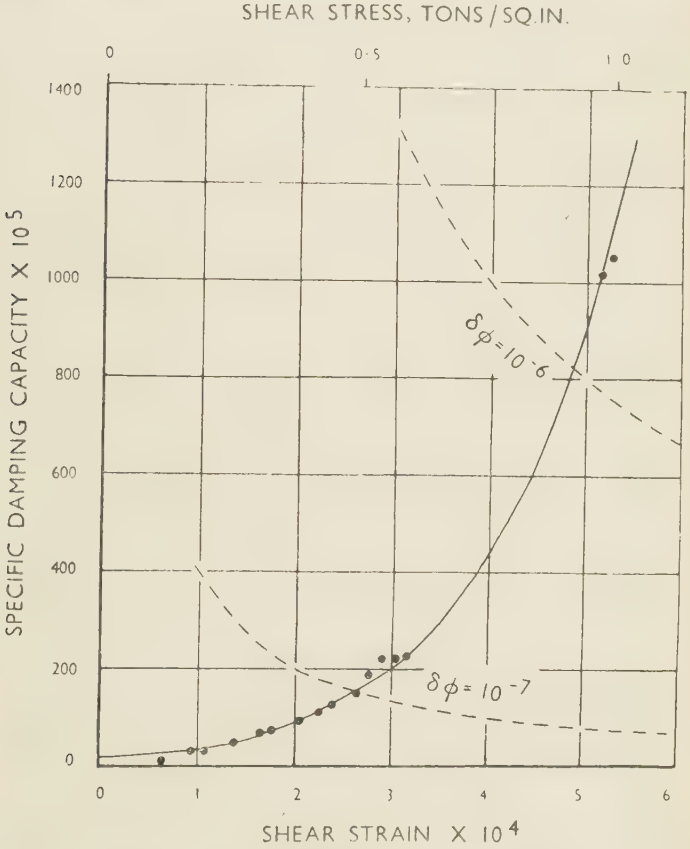

Fig. 1.—R.R. 56, Annealed at 360° C. (*An*-360).

in various metallurgical conditions. Vibration was not continued any longer than was necessary to make the experimental observations, because even short periods of vibration at high strain amplitudes were found to change the properties of the alloy.

The results obtained are shown in Figs. 1–6, the specific damping capacity being calculated from equation (5) after subtracting a small constant damping, approximately 5×10^{-5}, attributed to external damping and, possibly, a small thermo-elastic effect.

In the *An*–360, *ST–AN*, and *ST* 20*A*, *ST* 70*A*, and *ST* 100*A* conditions (Figs. 1, 6, and 5, respectively) the alloy was found to be comparatively stable with respect to time at room temperature, but in the *ST* 20, *ST* 70, and *ST* 100 conditions, the alloy changed in properties during ageing at room temperature. A series of tests was made on specimens in these conditions at intervals after quenching, and Figs. 2,

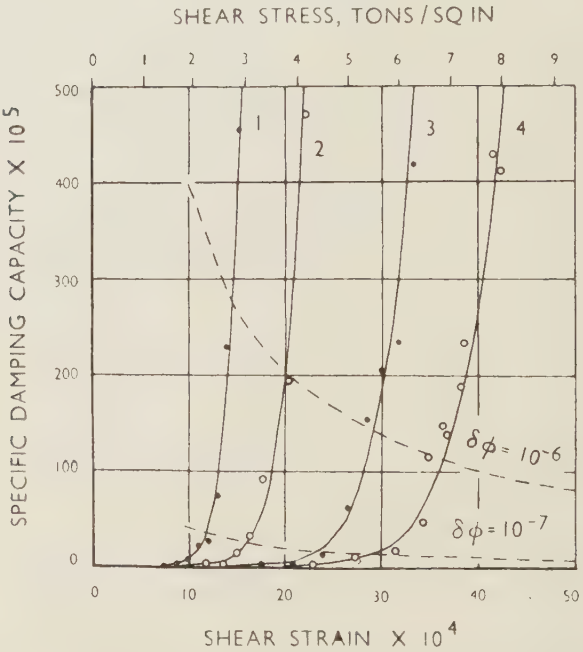

Fig. 2.—R.R. 56, Solution Heat-Treated and Quenched at 20° C. (*ST* 20).

Effect of time after quenching:

(1) immediately after quenching.
(2) 2·5 hr. ,, ,,
(3) 20 hr. ,, ,,
(4) 265 hr. ,, ,,

3, and 4 illustrate the magnitude of the effect of room-temperature ageing on the relation between damping capacity and strain.

The extent to which the observed variation of damping capacity with strain is represented by equation (4) is also indicated in Figs. 1–6, where the full curves are obtained from this equation, using values of the constants ϕ_m and q derived from experimental values of ϕ and $\Delta E/E$. It will be seen that in most cases the experimental points lie

on or near curves given by equation (4). The values of ϕ_m and q corresponding to these curves are given in Table I.

This first series of experiments shows that the variation of damping capacity with strain amplitude is of the general form given by equation

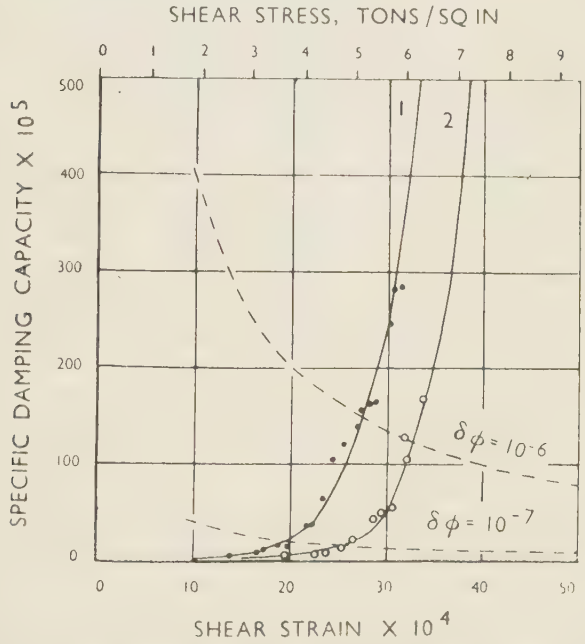

FIG. 3.—R.R. 56, Solution Heat-Treated and Quenched at 70° C. (*ST* 70).
Effect of time after quenching:
(1) immediately after quenching.
(2) 72 hr. ,, ,,

(4), the magnitude of the quantities ϕ_m and q being dependent on the metallurgical condition of the alloy.

3. *The Effect of Vibration.*

The second series of tests shows how the alloy, in various initial states, is affected by prolonged vibration. The factors governing torsional resonance vibration of the specimens are related by the equation:

$$\phi \,.\, \Delta E/E = C\tau \quad . \quad . \quad . \quad . \quad . \quad . \quad (6)$$

where ϕ is the surface shear-strain amplitude, $\Delta E/E$ is the damping capacity of the specimen, τ is the amplitude of the couple maintaining

TABLE I.—*Representation of Experimental Results by Equation* (4).

Condition of Alloy.	$q\phi_m$.	q.	$\phi_m \times 10^4$.	$G\phi_m$, tons/in.^{2}
An–360	0·385	206	18·7	3·52
ST–AN	0·494	100	49·4	9·28
ST 20 (immediately) . .	0·572	175	32·7	6·15
(after 2½ hr.) . .	0·572	127	45·0	8·45
(after 20 hr.) . .	0·572	82	69·8	13·1
(after 265 hr.) . .	0·572	64·5	88·8	16·7
ST 70 (immediately) . .	0·491	58	84·7	15·9
(after 72 hr.) . .	0·556	67·5	82·5	15·5
ST 100 (immediately) . .	0·481	85	56·6	10·6
(after 24 hr.) . .	0·507	66·7	76·0	14·3
(after 96 hr.) . .	0·529	65·8	80·3	15·1
ST 20*A* (after 3 days) . .	0·576	60	96·0	18·0
ST 70*A* (after 3 days) . .	0·590	69	85·5	16·1
ST 100*A* (after 3 days) . .	0·527	67·6	78·1	14·7

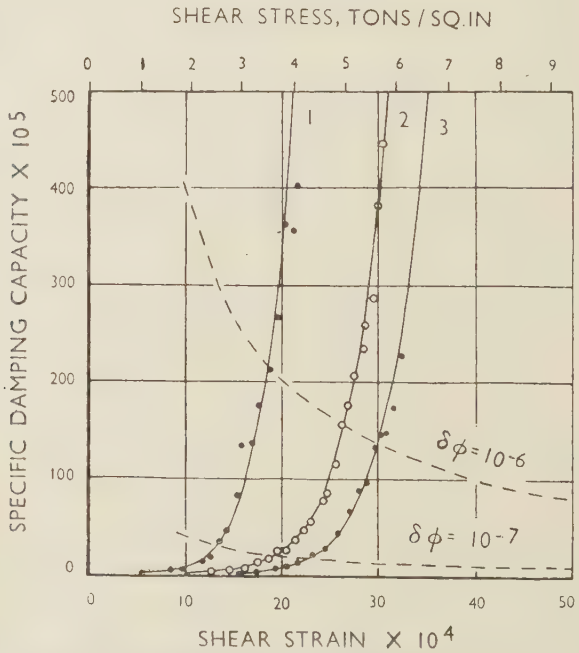

FIG. 4.—R.R. 56, Solution Heat-Treated and Quenched at 100° C. (*ST* 100).

Effect of time after quenching:

(1) immediately after quenching.
(2) 24 hr. ,, ,,
(3) 96 hr. ,, ,,

the vibration, and C is a constant calculated from the dimensions and rigidity modulus of the specimen.

If the magnitude of τ is held constant during vibration of the specimen (a condition which is easily established experimentally), any change in metallurgical condition, causing an alteration of the damping capacity, will be indicated by a change in the vibration amplitude. The magnitude

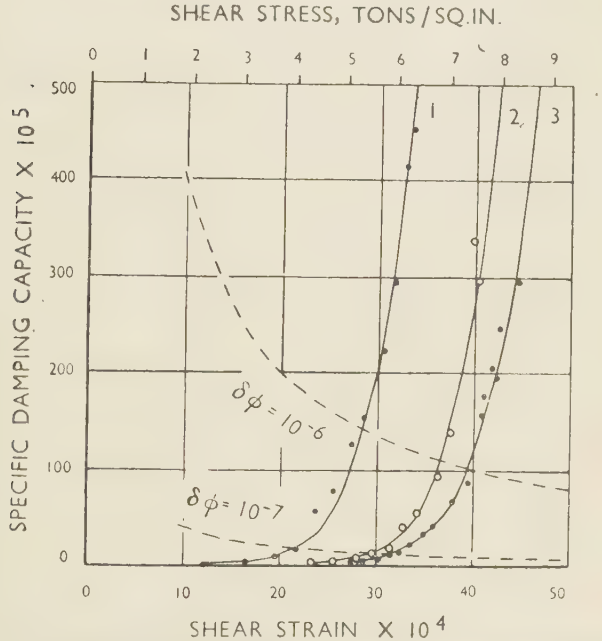

Fig. 5.—R.R. 56, Solution Heat-Treated, Quenched, and Aged at 170° C.

(1) Quench temperature 100° C (*ST* 100*A*).
(2) „ „ 70° C. (*ST* 70*A*).
(3) „ „ 20° C. (*ST* 20*A*).

Tested 3 days after completion of ageing at 170° C.

of τ may be found from the measurement of the damping capacity corresponding to a particular strain amplitude.

Figs. 7–10 show examples of the change in strain amplitude during prolonged vibration at various values of τ. These figures refer to the alloy in three typical states, namely *ST* 20 (Figs. 7 and 8), *ST* 20*A* (Fig. 9), and *ST–AN* (Fig. 10), but it is probably true to say that there is no state of the alloy which is absolutely stable during prolonged vibration. Even in the annealed state, very small changes in amplitude may be observed (see e.g. Fig. 11 of reference 5).

In order to differentiate between the effects of vibration and natural ageing of the alloy in the *ST* 20 condition, some specimens were vibrated, immediately after quenching, for periods not exceeding 3 hr. (approximately 10^7 cycles). The variation of amplitude during this period for different values of τ is shown in Fig. 7. For other specimens, vibration was commenced at various intervals after quenching, τ being

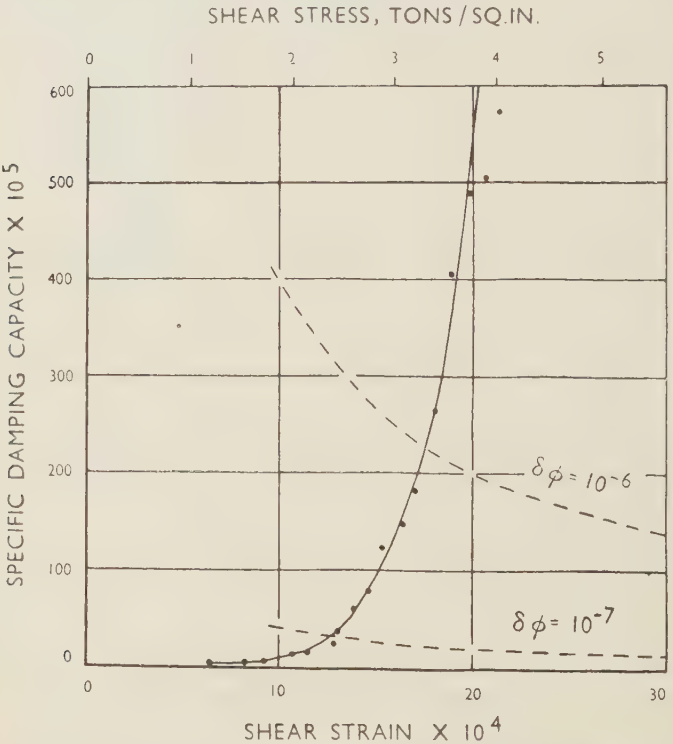

Fig. 6.—R.R. 56, Solution Heat-Treated, Cooled in Air, and Aged at 200° C. (*ST–AN*).

approximately the same in all the tests. Fig. 8 shows how the character of the vibrational instability of *ST* 20 gradually changes to that of *ST* 20*A* as the time between quenching and testing increases.

A general survey of the experimental observations recorded in Figs. 7–10 indicates that two forms of instability may be initiated in this alloy by vibration, one causing a decrease and another causing an increase in damping capacity. A decrease is apparently always followed eventually by an increase in damping, but for the alloy in certain

metallurgical conditions, an increase in damping capacity is the immediate result of vibration.

This phenomenon was investigated further, using alloy in the *ST* 20*A* condition, i.e. a condition in which the damping capacity increases from the commencement of vibration. Specimens vibrated at selected values

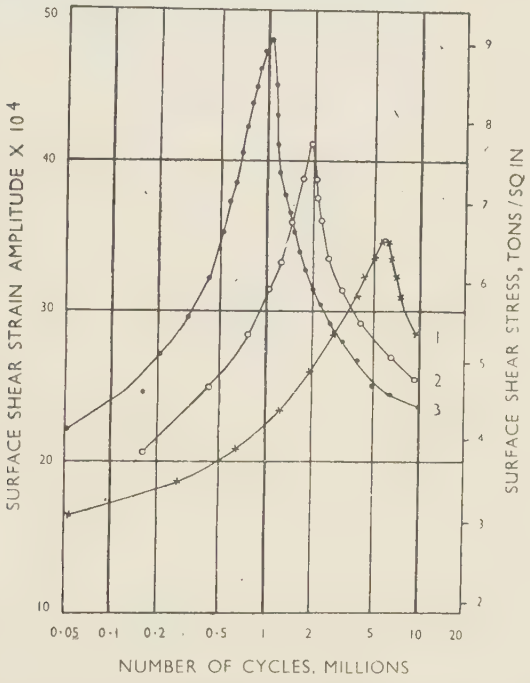

Fig. 7.—R.R. 56, Solution Heat-Treated and Quenched at 20° C. (*ST* 20). Effect of continuous vibration, commenced immediately after quenching:

(1) Vibrated at $\tau =$ 46 g.cm.
(2) „ „ $\tau =$ 92 g.cm.
(3) „ „ $\tau =$ 130 g.cm.

of τ were examined to determine the relation between strain amplitude and specific damping capacity after prolonged vibration. The results are shown in Fig 11, where the full curves are of the type given by equation (4). Curve 1 is obtained when the alloy is subjected to only the minimum amount of vibration necessary to allow the strain amplitude–damping capacity curve to be determined. Curve 2 represents the state of the alloy after vibration for 10^9 cycles by a small couple of amplitude 3·7 g.cm. The damping capacity has increased at all values

of strain amplitude, but the experimental points do not follow a curve of the type given by equation (4) when the strain amplitude exceeds 22×10^{-4}. After vibration for 10^9 cycles by couples of amplitude 38 and 78 g.cm. respectively, curves 3 and 4 were obtained, and these also show departure at high strain amplitudes from the type of curve given by equation (4). Curve 5 shows the final condition of another specimen after vibration by a still higher couple, 92 g.cm., but for a smaller number of cycles, namely $0{\cdot}35 \times 10^9$ cycles. Except at the highest

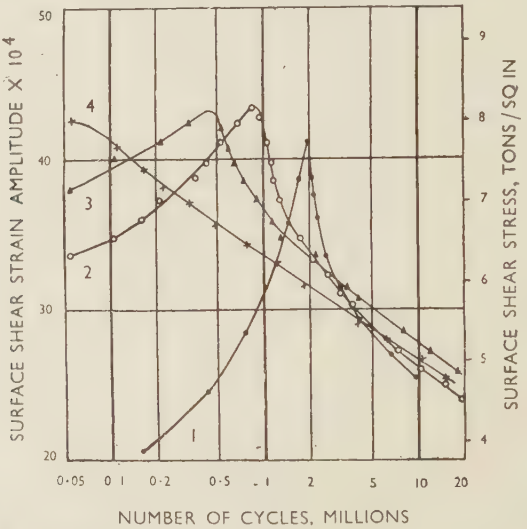

FIG. 8.—R.R. 56, Solution Heat-Treated and Quenched at 20° C. (*ST* 20).
Vibration commenced (1) immediately after quenching.
" " (2) 24 hr. " "
" " (3) 48 hr. " "
" " (4) 312 hr. " "
$\tau = 100$ g.cm. (approx.)

strains, the experimental points for this specimen lie on the same theoretical curve as those at the lower strains for the specimen vibrated for 10^9 cycles at $\tau = 78$ g.cm.

The values of ϕ_m and q corresponding to the full curves in Fig. 11 are given in Table II, which shows that q increases and ϕ_m decreases during vibration. It is apparent that if the vibration is maintained by a couple of high amplitude the alloy can be changed from the *ST* 20*A* condition to a condition (represented by curve *D*, Fig. 11) which approaches that obtained after annealing the alloy. Furthermore, after being changed from the *ST* 20*A* condition by prolonged vibration it

is found that a second heat-treatment to the *ST* 20*A* condition restores the variation of damping capacity with strain amplitude to the type characteristic of that condition. This observation is of importance

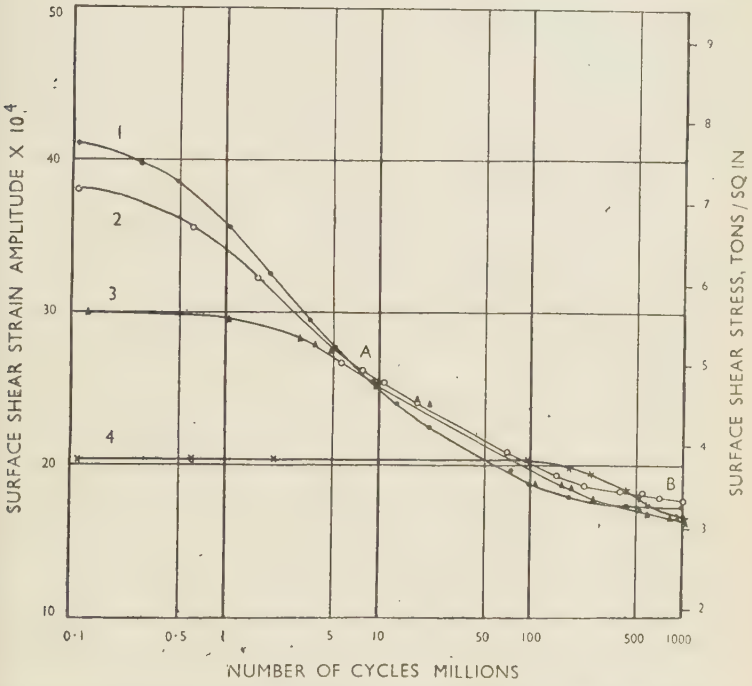

FIG. 9.—R.R. 56, Solution Heat-Treated, Quenched at 20° C., and Aged at 170° C. (*ST* 20*A*).

(1) Vibrated at $\tau = 78$ g.cm.
(2) ,, ,, $\tau = 37$ g.cm.
(3) ,, ,, $\tau = 9$ g.cm.
(4) ,, ,, $\tau = 4$ g.cm.

TABLE II.—*Values of Constants in Equation* (4) *for Vibrated R.R.* 56 (*ST* 20*A Condition*) *for Restricted Ranges of Strain.*

No. of Cycles.	τ, g.cm.	$q\phi_m$.	q.	$\phi_m \times 10^4$.	$G\phi_m$, tons/in.^{2}
Minimum . .	...	0·576	60	96	18·0
10^9 . . .	3·7	0·511	70·5	72·5	13·6
10^9 . . .	38	0·503	137	36·8	6·9
10^9 . . .	78	0·482	139	34·6	6·5
$0{\cdot}35 \times 10^9$ . .	92	0·482	139	34·6	6·5

because it shows that the transition of the damping-capacity curve from *A* to *D*, Fig. 11, during vibration, is not due to the formation of flaws within the alloy, unless indeed these flaws are on so minute a scale that they are healed by further heat-treatment.

As the effect of vibration on the relation between damping capacity and strain amplitude is quite large, experiments were made to see whether normal static torsion tests would show differences between vibrated and non-vibrated specimens. Specimens in state *A* and state

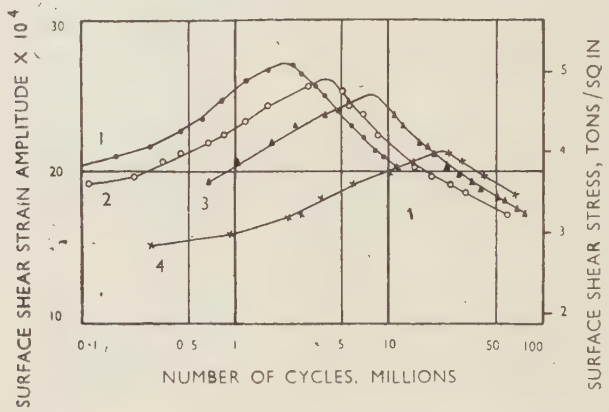

FIG. 10.—R.R. 56, Solution Heat-Treated, Cooled in Air, and Aged (*ST–AN*).

(1) Vibrated at $\tau = 127$ g.cm.
(2) „ „ $\tau = 80$ g.cm.
(3) „ „ $\tau = 40$ g.cm.
(4) „ „ $\tau = 10$ g.cm.

D (Fig. 11) were tested by straining torsionally until fracture occurred. As the specimens varied slightly in grain-size, although taken from one bar, they were examined as pairs, the specimens in each pair being those most comparable in grain-size. As shown in Table III, the stress-strain curves revealed no significant difference between the vibrated and non-vibrated specimens.

This failure of the static torsion test to show significant differences between vibrated and non-vibrated material may occur for the following reasons. During vibration, the strains are not uniform, varying from zero at the vibration axis to a maximum at the surface of the specimen. Consequently, vibration will affect the surface layers more than the central core. Experiment shows that if a surface layer, 1 mm. in thickness, is removed from a specimen vibrated to state *D*, Fig. 11, a large proportion of the material affected by vibration is removed. The specimen, reduced in diameter, then gives a relation between

TABLE III.—*Torsion Tests on Vibrated and Non-Vibrated Specimens.*

Static Shear Stress.	First Pair		Second Pair	
	not vibrated, tons/in.²	vibrated, tons/in.²	not vibrated, tons/in.²	vibrated, tons/in.²
0·01% Proof . . .	9·9	9·4	10·3	9·9
0·02% Proof . . .	11·0	10·4	11·1	11·0
0·05% Proof . . .	12·5	11·9	12·5	12·2
0·1% Proof	13·5	13·4	13·6	13·2
0·2% Proof	14·8	14·6	14·7	14·5
0·5% Proof	16·1	15·5	16·3	16·3
Ultimate	18·3	17·8	18·3	18·3

damping and strain amplitude which is approximately the same as curve *A*, Fig. 11, showing that the central core of the original specimen was not substantially affected by the vibrational strains prevailing in

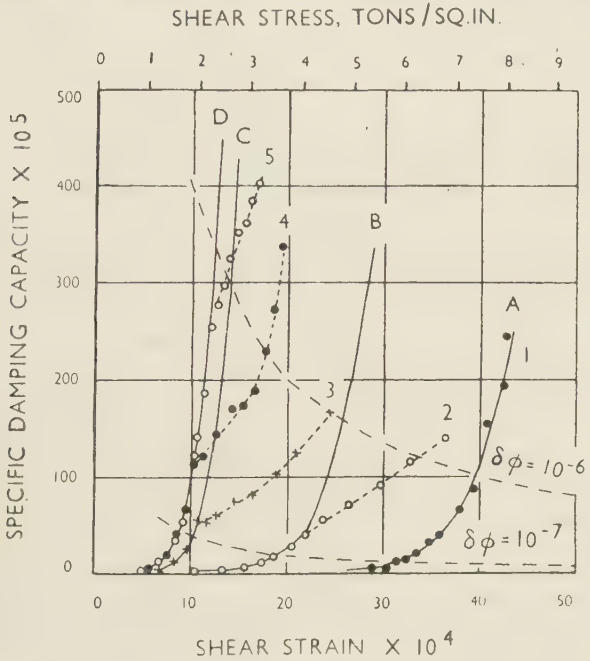

FIG. 11.—Effect of Vibration on R.R. 56 (*ST* 20*A*).

(1) Minimum vibration.
(2) After 10^9 cycles, $\tau =$ 3·7 g.cm.
(3) ,, ,, $\tau =$ 38 g.cm.
(4) ,, ,, $\tau =$ 78 g.cm.
(5) After $0{\cdot}35 \times 10^9$ cycles, $\tau =$ 92 g.cm.

this region. This central core will tend to mask the properties of the outer layer during the static torsion test. Furthermore, since the static test is sensitive to the effects of plastic deformation only when this deformation is considerable (e.g. a proof stress corresponding to less than 0·01% plastic deformation could not be determined accurately with the equipment available), the change produced by vibration may be masked by the effects of deformation produced subsequently during the static test. The possibility of detecting changes produced by vibration, by subsequent mechanical tests, might be greater if the specimen could be vibrated longitudinally, since, by this mode of vibration, the material would be affected uniformly over the cross-section.

IV.—Discussion.

A considerable proportion of this paper is necessarily devoted to the determination and analysis of the relation between damping capacity and strain amplitude before vibration produces significant changes in the alloy.

1. *Virtual Strength.*

It is found that the "instantaneous" state of the alloy can be represented by an equation containing two quantities ϕ_m and q having values depending on the heat-treatment. On theoretical grounds $q\phi_m$ is anticipated to be of the order of 0·5, and the experimental values are found to be of this order of magnitude. The strain-concentration factors q also appear to be of the correct order of magnitude, for they give values of $G\phi_m$ which approach the shear strength of the alloy. In equation (4), $G\phi_m$ represents the stress at which dislocations are generated without help from thermal fluctuations, i.e. a stress which should cause fracture provided that, during its application, the properties of the material are not changed, for example, by strain-hardening. The quantity $G\phi_m$ may, therefore, be regarded as the virtual strength [15] for a particular state of the alloy.

When the alloy is in the *ST* 20*A* condition, the virtual strength is 18·0 tons/in.^{2}, i.e. practically the same as the ultimate shear strength of 18·7 tons/in.^{2} found for the same material by a static shear test. For other conditions of the alloy, the virtual strength is generally less than the static shear strength, probably because the static test causes considerable strain-hardening before fracture occurs.

2. *Internal Strain and Limit of Proportionality.*

The values of $q\phi_m$ (Table II) obtained from the experimental data actually vary somewhat according to the state of the alloy.

It seems unlikely that the shear strain which is necessary to generate a dislocation will vary to this extent in materials which have essentially the same crystal structure. However, the possibility of the crystal lattice being in a state of strain, arising, for example, from precipitation centres or " arrested " dislocations must not be ignored. An internal field of strain may prevent slipping on glide planes until the applied stress is sufficient to overcome the internal strain. In effect, an internal strain may be indicated by an increase of $q\phi_m$ above its normal value. If it is assumed that $q\phi_m$ has a certain value Φ when there is no internal strain, then for a material having a value of $q\phi_m$ greater than Φ, the internal strain will be represented by:

$$q\phi_y = q\phi_m - \Phi \quad . \quad . \quad . \quad . \quad . \quad (7)$$

where ϕ_y represents the external strain which must be applied to overcome the internal strain. Thus $G\phi_y$ will represent the external stress which must be applied before dislocations can be generated and translated at an appreciable rate, i.e. it will represent the yield stress or limit of proportionality of the material. This limit cannot be measured precisely because it is impossible to establish experimentally that the damping capacity due to plasticity falls to zero at a finite strain. To obtain an estimate, the total strain corresponding to a small defined plastic strain may be adopted, as in the measurement of proof stresses. For a hysteresis loop approximating to a parallelogram:

$$\Delta E/E = 4\delta\phi/\phi \quad . \quad . \quad . \quad . \quad . \quad (8)$$

where $\delta\phi$ is the plastic component of the total strain ϕ. When a value of $\delta\phi$ is chosen, this equation is a hyperbola relating all values of ϕ and $\Delta E/E$ corresponding to a proof stress defined as the stress producing the plastic strain $\delta\phi$. For example, the broken curves in Figs. 1–6 and Fig. 11 are for $\delta\phi = 10^{-6}$ and $\delta\phi = 10^{-7}$, and the strains at which these curves intersect the experimental curves relating damping capacity and strain give respectively the strains corresponding to 0·0001% and 0·00001% proof stresses. However, it is clear that a less objectionable procedure is to estimate the limit of proportionality by measuring the strains for which the plastic component bears a definite ratio to the elastic component, i.e. strains for which $\Delta E/E$ has a small defined value. In the present work the error involved in measuring $\Delta E/E$ is not less than 10^{-5}, so that $\Delta E/E = 10^{-4}$ is a reasonable value to adopt for estimation of the limit of proportionality. For this value, the ratio of plastic to elastic strain will be approximately $2{\cdot}5 \times 10^{-5}$.

Table IV give values of ϕ_y and $G\phi_y$ estimated on this basis from the experimental data given in Figs. 1–6. The final column shows the

corresponding values of Φ calculated from equation (7) by employing the values of q and ϕ_m given in Table I.

TABLE IV.—*Limits of Proportionality and Calculated Values of* Φ.

Condition of Alloy.	$\phi_y \times 10^4$.	$G\phi_y$, tons/in.^{2}	$q\phi_y$.	Φ.
An–360	0·0	0·0	0·0	0·385
ST–*AN*	10·2	1·91	0·102	0·392
ST 20 (immediately) . .	10·2	1·91	0·178	0·384
(after 2½ hr.) . .	14·6	2·74	0·185	0·387
(after 20 hr.) . .	21·8	4·10	0·179	0·393
(after 265 hr.) . .	27·0	5·05	0·174	0·398
ST 70 (immediately) . .	17·0	3·19	0·099	0·392
(after 72 hr.) . .	22·5	4·21	0·152	0·404
ST 100 (immediately) . .	11·0	2·06	0·093	0·388
(after 24 hr.) . .	17·0	3·19	0·113	0·394
(after 96 hr.) . .	21·0	3·94	0·138	0·391
ST 20*A* (after 3 days) . .	31·2	5·85	0·187	0·389
ST 70*A* (after 3 days) . .	28·5	5·35	0·197	0·393
ST 100*A* (after 3 days) . .	19·4	3·64	0·131	0·396

There appears to be little doubt that Φ is constant, and therefore it is concluded that when the crystal structure is free from internal strain, the strain necessary to cause an atom jump which will generate a dislocation is approximately 0·39. The magnitude of the internal strain of the aluminium lattice, probably originating from foreign atoms in a state of solution or dispersion, is evidently given by $q\phi_y$ and, as might be expected, $q\phi_y$ is large when the alloy is severely quenched after solution heat-treatment.

3. *Strain-Concentration Factor.*

The significance of the strain-concentration factor q is explained on the basis of a suggestion by Taylor.[10] In any crystal grain there are weak regions, in an otherwise strong structure. When a shearing stress is applied, the majority of the shear strain occurs in the weak regions, with very little strain in the intermediate zones. Suppose that the externally applied strain is ϕ_e and that the weak regions, having a width x small compared with their spacing y, develop a strain ϕ. Then:

$$q = \phi/\phi_e = 1 + y/x = y/x \quad . \quad . \quad . \quad . \quad . \quad (9)$$

approximately. As y will be an inverse function of the number of weak regions, a decrease in q will indicate an increase in the number of

such regions. The actual strength of the material will increase as the number of weak regions increases, because then the strain concentration in each region will decrease.

Table I shows that immediately after quenching the alloy at 20° C., q is large, but decreases with time at normal room temperature or after ageing at elevated temperature. Quenching at a higher temperature gives values of q, immediately after quenching, which are smaller than the value of q after quenching at 20° C. Thus, in general, q decreases as precipitation proceeds towards a state giving maximum mechanical strength. It seems that the regions where strain concentrations occur may be associated with the products of precipitation, possibly the plate-like collections of atoms observed by Guinier and by Preston.[16]

If the alloy is annealed, the precipitation centres may be expected to grow in size but to decrease in number, i.e. their spacing will increase. Thus, q may be expected to be increased by annealing, and Table I shows that this is so.

The strength of the alloy is also determined in part by the internal strain $q\phi_y$, and it should be noted that this is not necessarily high when the precipitate has a distribution favourable to strength (e.g. *ST* 70, immediately after quenching). The internal strain may be increased by slow change in the precipitate during ageing.

4. *Effect of Vibration.*

The preliminary effect of vibration on solution heat-treated and quenched alloy (*ST* 20) is shown, in Figs. 7 and 8, to be equivalent to an acceleration of the process of natural ageing. Prolonged vibration causes the precipitation process to continue. Thus, if the alloy is brought to a condition in which it has high strength (e.g. peak of the curves in Figs. 7 and 10 by vibration; peak of the curves in Fig. 8 by natural ageing; initial state in Fig. 9 by elevated-temperature ageing), further vibration causes growth and coalescence of precipitation centres, leading to an increase of the strain-concentration factor and consequently a decrease in strength. In a resonant system this increase of q causes an increase in the damping capacity, and the strain amplitude consequently decreases if the amplitude of the couple causing the vibration remains constant. Also, in such a system the increase in damping capacity causes a *decrease* in the amount of energy dissipated by the alloy. Thus, as vibration proceeds, under constant torque the changes caused in the state of the precipitate are likely to occur at a decreasing rate.

While it is true that the general effect of vibration is similar to that of temperature, differences between the action of these stimulants of

precipitation become evident on closer inspection. For example, Fig. 2 represents change in the alloy due to ageing at room temperature, and it is seen that the alloy progresses through states represented by curves of the type given by equation (4). In Fig. 11, representing changes caused by vibration, the final state D is not approached through a succession of states expressed by this equation.

This difference is explained as follows. Thermal fluctuations may affect any crystal grain, whereas mechanical vibration, being directional, may be selective in its action, more grains being influenced when the amplitude of the applied couple is increased. If a specimen is vibrated for a long time by a couple having an amplitude τ, certain grains will be affected. Suppose that the specimen is then tested to determine the variation of damping capacity with strain amplitude, by taking observations at values of τ increasing by stages from zero. The grains which will determine the characteristics of the specimen will be, until τ_1 is reached, those grains affected by the previous prolonged vibration. When τ_1 is exceeded, new grains, not seriously affected by the prolonged vibration, will contribute more and more effectively to the damping capacity. Thus, a break in the smooth curve relating damping capacity to strain amplitude may be expected when the couple τ_1 is exceeded. Breaks of this type appear in Fig. 11, and if this is a true explanation for their existence it provides a further reason why the gross mechanical properties, as determined by static tests, do not show conclusive differences between the vibrated and non-vibrated alloy. The grains affected by vibration will be, nevertheless, those which determine the fatigue properties of the alloy.

5. *Fatigue.*

Interpretation of the data presented in this paper in terms of the fatigue resistance of the alloy is a matter for inference, for, out of a large number of specimens examined, only one developed a fatigue crack during vibration. It was evident that this failure occurred as a result of a material defect.

Referring to Fig. 9, it will be seen that after a few million cycles, the strain amplitude after a given number of cycles is approximately the same for a wide range of high values of τ. Equation (6) shows that the energy dissipated per cycle is then nearly proportional to τ. Consequently, the energy dissipated per cycle by the alloy increases very greatly for very small increases in strain amplitude. At some high rate of dissipation of energy, fatigue failure will occur, but the strain amplitude at failure will not be appreciably higher than the strains shown in Fig. 9 over the section AB of the curves. The section AB is

therefore an approximate relation between strain amplitude and number of cycles for safe conditions of vibration.

Table V shows the shear stresses likely to occur in R.R. 56 as a result of resonance vibrations excited by a couple of high amplitude, the external losses of vibrational energy being small. It is the change produced in the alloy during vibration which causes the magnitude of these stresses to decrease as vibration proceeds. If, however, the alloy were to be forced to vibrate at a constant stress amplitude, say 5·25 tons/in.^{2} shear, it is likely that it would fail by fatigue soon after 5×10^6 cycles, and therefore the stresses shown in Table V may be regarded as stresses which it is safe to apply for the stated number of cycles.

TABLE V.—*Magnitude of Resonance Stresses Likely to Occur in Solution Heat-Treated and Aged R.R.* 56 *During the Course of Prolonged Vibration.*

Number of Cycles, millions.	Shear-Strain Amplitude, $\times 10^4$.	Shear-Stress Amplitude, tons/in.^{2}
5	28	5·25
10	25	4·70
20	22·8	4·28
50	20·4	3·84
100	19·0	3·57
200	17·8	3·34
500	17·3	3·25
1000	17·2	3·23

It is interesting to compare the vibrational behaviour of this alloy with that of the alloys aluminium–5% magnesium and aluminium–11% magnesium previously investigated.[6] During resonance vibration, if the initial stress amplitude exceeds 2·8 tons/in.^{2} shear for aluminium–5% magnesium or 5·6 tons/in.^{2} shear for aluminium–11% magnesium, these alloys strain-harden continuously for a constant value of τ, until fatigue occurs at stress amplitudes above 6·1 tons/in.^{2} for aluminium–5% magnesium, or 9·0 tons/in.^{2} for aluminium–11% magnesium. Although these fatigue failures occur at comparatively high stress amplitudes, the value of τ required to produce these amplitudes in a system having low external damping is small. For example, specimens of these alloys inevitably failed by fatigue in about 150 million cycles or less if τ was greater than 8 g.cm.

In contrast to this behaviour, R.R. 56 in the fully heat-treated state does not fail in less than 1000 million cycles for values of τ at least up to 90 g.cm. Even when R.R. 56 is initially in a state (i.e. solution heat-treated and quenched) which strain-hardens during vibration, the strain-

hardening does not continue indefinitely until fracture occurs, but stops when the alloy is brought to a state comparable with that produced by ageing at 170° C.

The probability of a fatigue failure of R.R. 56 at resonance under conditions where the external damping is small is therefore very much less than the probability of failure of the aluminium–magnesium type of alloy.

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